

THE CASTING OF BRASS INGOTS

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LONDON :
**BRITISH NON-FERROUS METALS
RESEARCH ASSOCIATION**

1934

Reprinted August 1943

PRINTED BY THE REPLIKA PROCESS
IN GREAT BRITAIN BY
LUND HUMPHRIES
LONDON · BRADFORD

INTRODUCTION

THE casting of the ingot is a vital stage in the manufacture of wrought metals, for although improvements in quality and economies in working may be effected in later operations, perfection cannot be achieved if the original ingot is faulty. The publication of a book concerned with the production of good ingots of non-ferrous metals, a subject which has been comparatively neglected until recently, is therefore to be welcomed. The bulk of the work described has been carried out on alloys of one type, ductile brasses used in the rolled condition, but these materials extend over a range of composition, are produced in large quantity and applied in an immense variety of objects in common use, and are thus highly important industrial products. It might at one time have been thought that the production of a cast rectangular brass slab free from all imperfections is a simple matter, but such disturbing factors as the rapid oxidation of the molten metal surface when exposed to air and the consequent formation of a troublesome oxide skin, the turbulence of the stream of metal in the mould, the speed of crystallisation and the contraction in volume on solidification—factors which are common to the casting of most industrial metals and alloys—unavoidably introduce defects into the ingot. Remedies for these troubles are not hard to find, but, as in so many of the affairs of life, measures taken to remove one difficulty are only too apt to give rise to others possibly still more objectionable. Again, the economic factor can never be neglected by the manufacturer, laudably intent on producing the best article at the price of the cheapest, and the cost of improvements must be rigorously scrutinised.

Like the five weighty reports already published by the Heterogeneity of Steel Ingots Committee, whose work is still continuing, this book is evidence of the increased interest now taken in the quality of ingots intended for working into rolled

or forged products. This quality is controlled, not only by the characteristics of the liquid metal but also by the whole sequence of casting and cooling. In spite of the great difference in the scale of operations, striking comparisons and contrasts will occur to readers who are also acquainted with the casting of steel. It will indeed be found that the results of the work described in this book are by no means limited to the alloys used, but have a much wider scientific and practical value in their relation to casting problems in general, both within and without the field of non-ferrous metallurgy.

The writer has had the privilege of being closely concerned with the experimental work forming the main substance of this book. He would like to pay a tribute to the skill and resource of the authors in tackling a long series of complex problems, no aspect of which has been neglected. The thanks of the particular industries concerned are now due to them for bringing their work and other current developments into a true perspective in the present review, which should also make a strong appeal to all others interested in the absorbing problems of metal casting.

H. MOORE.

PREFACE

THE practice of brass-making, although in the aggregate an important metallurgical industry, has been until recently carried out almost entirely in numerous small foundries, each separately dependent on traditionally trained foundry workers and in the majority of cases without the resources necessary for technical research. Under these conditions no radical changes of method occurred during the industrial development of the nineteenth century, and manufacture remained largely controlled by a technique built up entirely by empirical means. The practical skill of the average brass caster and the care which he exercised in his work have been reflected in the general high standard of quality of brass sheet and strip produced. In the absence of scientific control, however, the causes of the common types of defect leading to difficulties in manufacture and the directions in which remedial measures might be sought could not be clearly ascertained. Establishment of the necessary knowledge of the process was essentially a matter requiring combined effort, as well as conditions which would permit of a collective survey of the many local variations in methods of casting. A channel for such co-operation was provided by the inception of the British Non-Ferrous Metals Research Association in 1920, and the casting of strip ingots of brass was one of the first subjects selected for investigation.

The object of the research was essentially the improvement of the quality of brass strip, to be reached mainly by the investigation of defects, the detailed study of the behaviour of brass in casting, and the development of improved methods. Although the cost of laboratory research is increased by working on a relatively large scale, it was considered that for a subject such as the casting of brass the disadvantages would be more than counterbalanced by the closer applicability of the results to manufacture. Since brass of 70/30 composition was not only the most important but also by general opinion one of the most

difficult common alloys to manufacture in high quality strip, most of the experimental work was concerned with this composition. The many factors affecting surface quality and soundness provided a sufficiently large field for the investigation and such aspects as mechanical properties were not included.

The research work, which was throughout maintained in close contact with manufacturers, was carried out by the authors and their colleagues at the Research Department, Woolwich, and since no previous work of the kind had been attempted systematically, the planning of the investigation was free from any limitations of scope. From a preliminary but thorough survey of works practice, it was apparent that a large number of factors were involved in the casting process, and accordingly the elucidation of the manner in which each of these separately acted was the course initially followed. The results of this work and of a number of related investigations arising from it were communicated to the industrial members of the Association from time to time during the progress of the research as reports and lectures. A list of these is given in Appendix E.

The present volume presents the results of the research in summarised form, amplified by discussion of their practical bearing, and in order to present a picture of the development of the process of brass casting leading to the recent work an introductory historical chapter has been compiled. In addition, a brief account of rolling practice has been included, and in the appendix dealing with the constitution of the copper-zinc alloys, advantage has been taken of the existence of data from other sources. Otherwise the matter given is entirely the product of the research carried out at Woolwich, at the laboratories of the British Non-Ferrous Metals Research Association, and in the works of manufacturers. During the progress of the work, however, important developments in melting and casting practice took place elsewhere, and suitable reference to these has been made. The space available precluded any attempt to prepare a complete treatise on brass or to deal with any extension of the subject-matter to other important aspects. It is the hope of the authors that the work will interest both manufacturers and users by making clear

some of the causes of defects in rolled brass and the principles to be observed in avoiding them. While much further work remains to be done in following up the general conclusions reached, the manner in which the quality of rolled brass is influenced by the various factors in the casting process may now be regarded as understood. Although intentionally and properly confined to one type of alloy, the work clearly applies to a much wider field, and the results may therefore be of direct interest to those engaged in the manufacture of ingots of other alloys where the difficulties are similar to those connected with the brass industry.

Briefly, and very inadequately, the authors take this opportunity of acknowledging their indebtedness in many directions to Dr. H. Moore, C.B.E. (formerly Director of Metallurgical Research, Woolwich), and Professor R. S. Hutton, D.Sc. (formerly Director of the British Non-Ferrous Metals Research Association); to their colleagues, Dr. L. Northcott for valuable assistance in experimental work and Mr. H. Wrighton, B.Met., F.R.M.S., for the photographic and photomicrographic illustrations. They would also express their appreciation of the helpfulness of the members of Research Sub-Committee No. 2 of the Association, who, under the chairmanship of Dr. H. W. Brownsdon, generously devoted much time to assisting the progress of the research and formed a valuable liaison with industry. In connection with the preparation of the historical chapter of the book, the authors wish to thank Mr. B. Fullman, B.Sc., for the collection and critical examination of much early literature, and also the Swedenborg Society, which kindly placed at their disposal a translation of a part of Swedenborg's "Opera Philosophica et Mineralia."

A number of the illustrations in this book are reproduced from blocks kindly supplied by the Institute of Metals, to whom the authors also wish to express their indebtedness.

R. GENDERS.

G. L. BAILEY.

London, January, 1934.

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CHAPTER I

THE HISTORY OF BRASS AND THE DEVELOPMENT OF BRASS INGOT CASTING

Earliest References—The Calamine Process of Brass Manufacture—
Ingot Casting to the Eighteenth Century.

THE initial use of metals by man was governed by their natural occurrence in the free state. The distribution of native copper was much wider in early days than is now the case, and early man made use of the metal for primitive tools and weapons by hammering small lumps into shape with stone hammers. Copper articles fashioned in this manner have been found dating back to 4500 B.C.,¹ and a long period of such use undoubtedly preceded the true smelting and casting of the metal. The discovery that copper could be melted in the fire and the later and more important discovery that it could be smelted from certain rocks by heating were probably accidental, and may have been made independently by a number of different people widely spaced in time and geographical situation. In such different districts the ores available were markedly different in composition, and the impurities present found their way to a certain extent into the copper. Various alloys were thus accidentally produced: Egyptian copper in general contained arsenic; German copper contained nickel; Indian copper contained zinc; Hungarian copper contained antimony, etc. It happened that tin was associated with copper in many districts and the production therefore of low tin-bronzes was not uncommon in early copper smelting.

Like bronze, brass can be numbered with the ancient alloys, and was undoubtedly made long before the second metal of the alloy was identified. The date of the first intentional production of brass is unknown and is made more uncertain by the misleading use of the word "brass" in ancient literature such as the Bible, where copper, bronze and possibly brass were all

described as "brass." About the third century B.C.² the Mossynaeci, a people living on the shores of the Black Sea, were producing brass by smelting with their copper an earth called "calmia" which was presumably calamine. Writing later, Dioscorides and Pliny both referred to this substance as "cadmia," and many centuries later this word persists in the descriptions of brass manufacture. There seems no doubt, however, that calmia or calamine is meant. Galen, for instance, in A.D. 166, describes as "cadmia" a material that is assuredly zinc oxide.³

Apart from the deliberate use of calamine a certain amount of brass was smelted direct from naturally occurring mixed ores of copper and zinc. Such "brass" mines existed in parts of India (notably Bidar in Hyderabad) and France, but the European supply was rapidly exhausted.

The manufacture of brass by empirical methods similar to that of the Mossynaeci was developed to a large extent by the Romans, who used the alloy for coinage and for decorative purposes. The following description of the Roman method is given by Gowland : ⁴

"The calamine was ground and mixed in suitable proportions with charcoal and copper in granules or small fragments. This mixture was placed in a crucible, and was very carefully heated for some time to a temperature sufficient to reduce the zinc in the ore to the metallic state, but not to melt the copper. The zinc being volatile, its vapour permeated the fragments of copper, converting them into brass. The temperature was then raised, when the brass melted and was poured out of the crucible into moulds."

The same method is described by Theophilus ⁵ in the eleventh century A.D. Theophilus also described at some length the methods used for refining the copper, to the purity of which some importance was attached.

For many centuries the calamine method was used substantially as described above. Metallic zinc, though known earlier in other parts of the world, and in fact discovered in certain bracelets found in the ruins of Kameiros, a city in the island of Rhodes which was destroyed in the sixth century B.C.,⁶ was not recognised in Europe until 1509, when Erasmus

Ebener of Nurnberg separated it from calamine.⁷ Ebener showed that brass could be produced by mixing zinc and copper, but he had only a small quantity of the metal available. In later years, however, although zinc became available in larger quantities, its adoption for the direct production of brass was still long delayed. It was not until 1781 that a patent was taken out in England by James Emerson¹⁵ for making brass with copper and spelter, and it needed almost another century for this direct method to become universal. Calamine brass was generally believed to be superior in mechanical properties to brass made by using metallic zinc. It is difficult to see the justification for this belief unless the spelter available was very impure, and conservatism was possibly responsible for the retention of the calamine process. Whatever the cause, it seems that this process was still in use in Birmingham in the middle of the nineteenth century. Direct alloying of the metals was practised on the Continent for certain special purposes prior to the date of Emerson's patent, though the calamine process was generally used. Swedenborg, writing in 1734,⁸ describes the production of brass and refers under this heading only to the calamine process. He states that scrap brass must be added to the charge to give the metal a good colour, and the pot filled up on top of this with alternate layers of "cadmia" (calamine?) mixed with charcoal dust and broken pieces of copper; 46 lbs. of cadmia should be mixed with 30 lbs. of copper and 20 or 30 lbs. of brass and the crucible heated for about fourteen hours. In the same work, Swedenborg also describes the production of "spurious" gold from copper and zinc, but no indication is given as to whether the essential similarity of this material and calamine brass was recognised :

"If zinc is mixed with copper a golden-coloured metal is produced, its colour varying according to the proportion of the mixture. If the copper be melted by fire and zinc be thrown into the melted mass, the zinc instantly takes to flight and is lost, but if the zinc be fused first and the copper put in, then both become amalgamated into the above-named metal, *i.e.* Prince Rupert's Metal."

As has been stated above, the first metallurgical furnace was the camp fire, and modern melting methods have developed from this by successive modifications. As a first step a shallow

cavity was formed in the hearth of the fire in which the molten metal was allowed to collect and solidify as a block about 8 to 10 inches in diameter. This was taken out when it had just become solid and broken up or worked to shape by hammering. Later, forced draught was obtained by utilising the winds, and this was followed by the introduction of primitive bellows, illustrations of which have been found on an Egyptian tomb of the fifteenth century B.C.⁹ Clay was used to line the hearth in order to give a smooth surface, and later this material was employed to make pots or crucibles. This development permitted manipulation of the molten metal, and Petrie¹⁰ records the casting of thick copper sheets in open sand moulds by the Egyptians. Ingot-shaped cavities were later cut in slabs of stone, and refractory materials of this type were used for mould construction until the introduction of metal moulds in comparatively recent times. An interesting exception to this practice was provided by the Japanese, who, according to Gowland,¹¹ cast plates of copper in canvas moulds submerged in a tank of hot water.

The manufacture of brass strip ingots was first described in the literature by Lazarus Ercker in 1574.¹² In the third part of his book, dealing with copper, he describes "how copper is to be made brass" by the calamine process. The method is substantially the Roman process described earlier, except that alum and salt were added to the calamine and coal mixture. After melting in eight small pots the brass was transferred to a large crucible and cast. Ercker adds, "but if you wish to make kettles, etc., then cast the contents of the pot into large stones made specially for the purpose, called Britannish stone (because they are brought from there), to form large plates. These can then be cut, drawn to wire or beaten as desired." A number of later editions of Ercker's work were produced and the illustration shown in Fig. 1 first appeared in the 1580 edition, but no fuller information than that which has already been quoted appeared in the description of this illustration.

A much fuller account of European practice was given by Swedenborg in 1734 (*loc. cit.*). In Swedish practice at this time the mixture of calamine with copper, etc., was heated in eight small crucibles in one furnace, the temperature being controlled by regulating the air supply. The crucibles and furnace linings

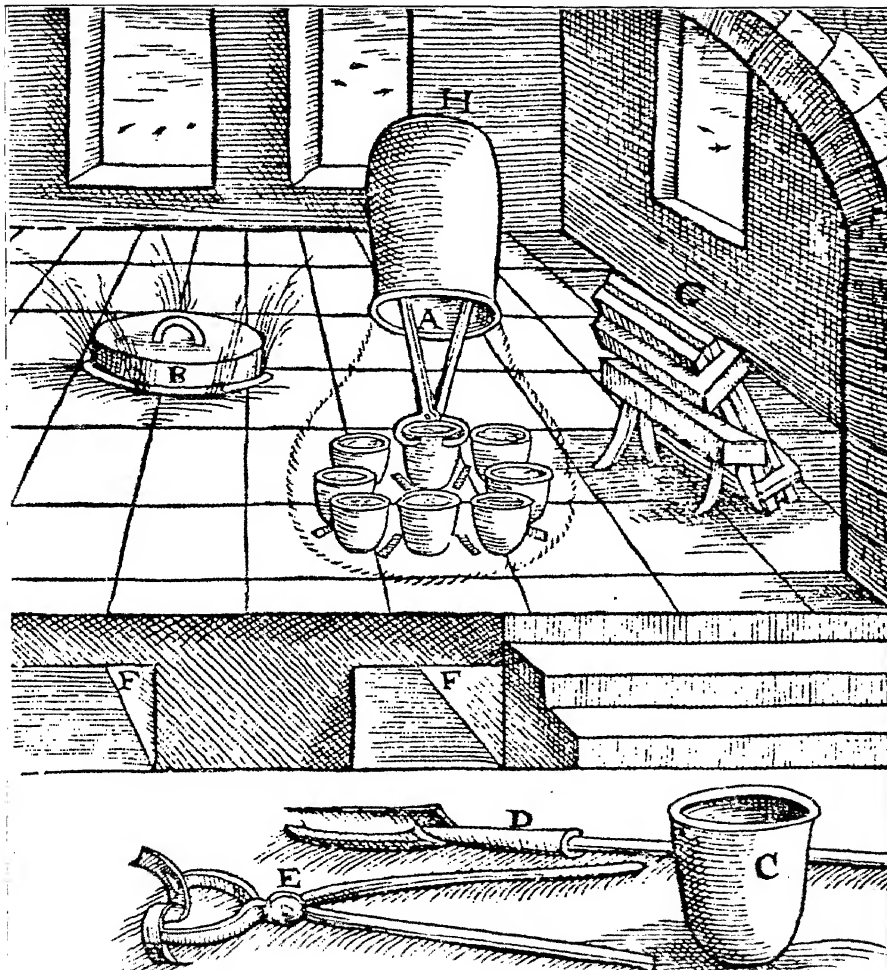
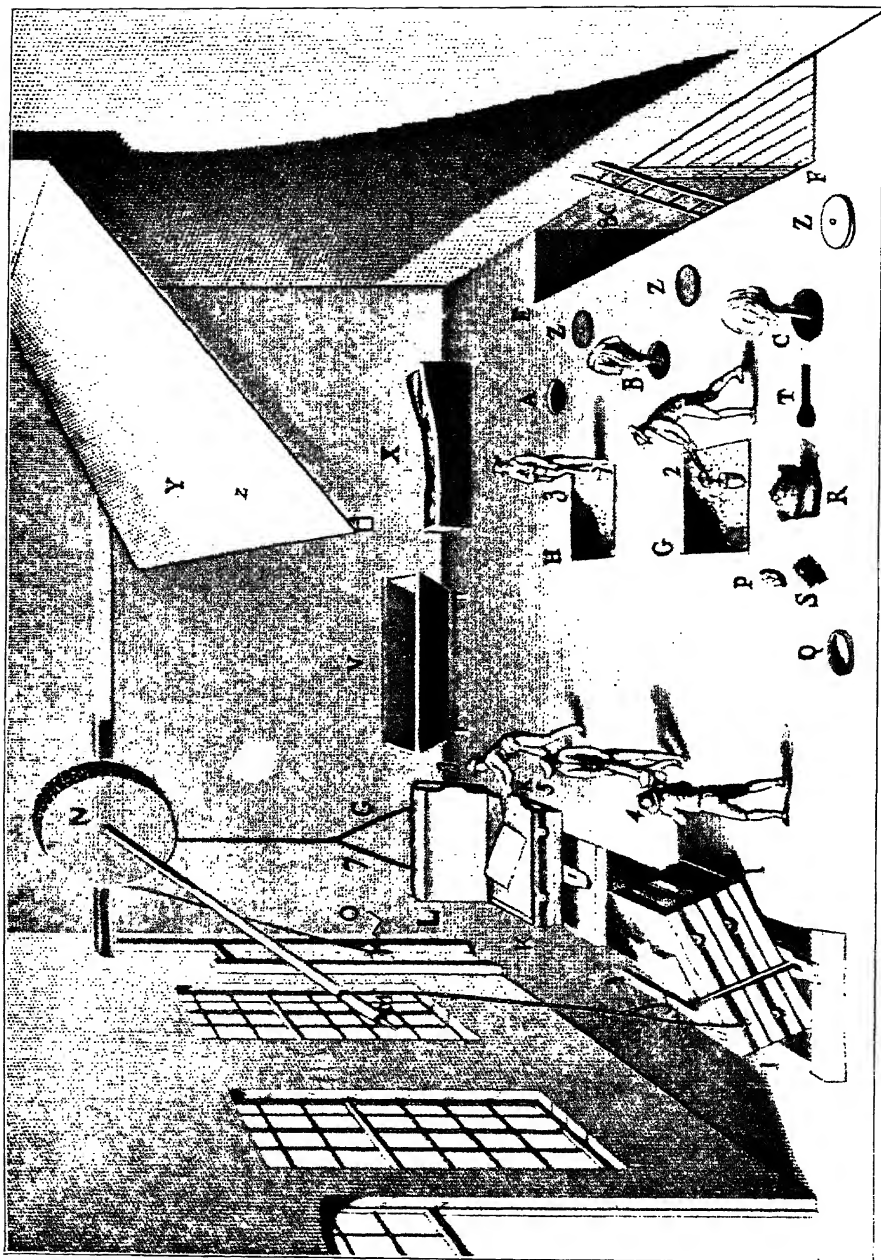


Fig. 1.

Diagrammatic illustration of brass foundry, published in 1580. (Ereker ¹².)

- A. Arrangement of crucibles in the furnace.
- B. The furnace in operation.
- C. Shape of crucibles.
- D. Shovel for calamine.
- E. Tongs for crucibles.
- F. Draught holes.
- G. Mould of "Brittannish stone."
- H. The master crucible into which the small crucibles are emptied.



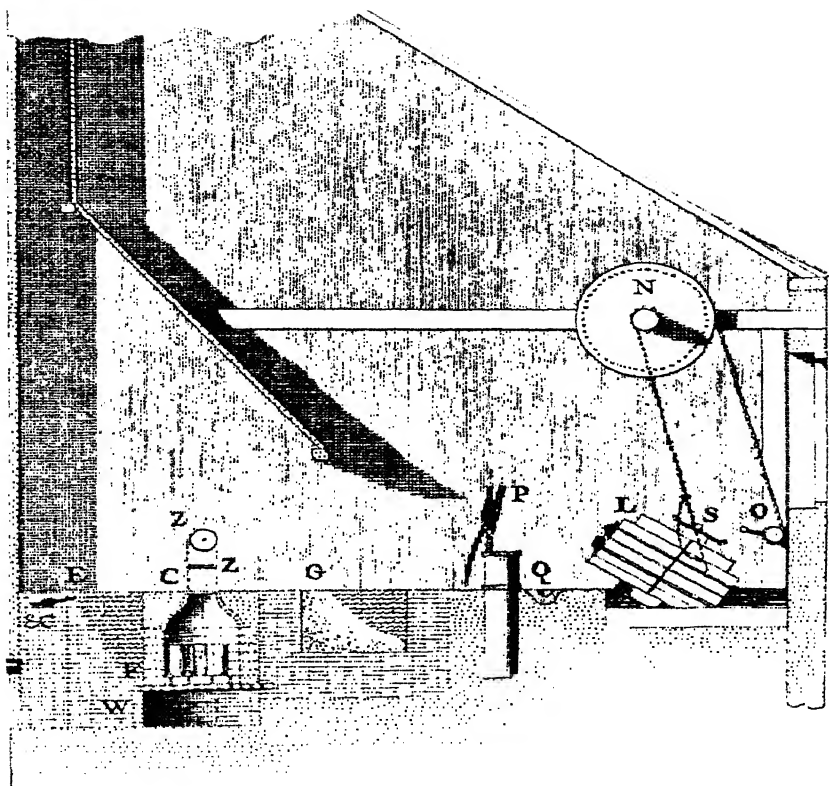


FIG. 3.
Section of foundry shown in Fig. 2.

Reference letters in Figs. 2 and 3 :

- A, B and C. Furnaces built in brickwork E and F.
- G and H. Pits into which crucibles are skimmed.
- I, K and L. Stone moulds held together with iron frames.
- M, N and O. Mechanical apparatus for opening moulds.
- P (Fig. 2). Ball of scrap metal.
- P (Fig. 3). Shears for cutting copper.
- Q. Mortar for the balling of scrap.
- R. Tub for measuring calamine.
- S (Fig. 2). Small copper cubes.
- T. Paddle for mixing the calamine.
- V. Tub in which the calamine and charcoal are mixed.
- W. Ashpit.
- X. Bed for foundrymen who stayed in the foundry for five full days of twenty-four hours each per week.
- Y. Apron for catching fumes.
- Z. Cover for furnaces.

were made of Welsh clay, and crucibles lasted for eight to ten weeks if handled carefully. In the centre of the fire was an empty crucible to which the brass from the eight small crucibles was transferred when ready. The brass "thus collected is poured out between two masses of sandstone so as to form a thick plate. The stones between which this melted mass of brass is poured are 5 feet long, $2\frac{1}{2}$ feet broad and 9 inches thick. They are held together by very thick iron bands and are raised by means of pulleys: they are also covered with a coarse woollen cloth. These stones are overlaid with clay." The moulds in question were apparently sandstone obtained from France. The use of metal moulds had evidently been considered at this time, as Swedenborg goes on to say, "an attempt has been made to pour out the smelted brass between iron plates, but as these plates cannot be overlaid with clay they are unsuitable, the stream of brass cannot penetrate as far as the other end of the space they enclose, but curdles and stops short, besides being full of holes. Stones which are of good quality may last four or five years, inferior quality three to five months." Swedenborg also gives a description of the making of brass at the Baptist Mills near Bristol. The methods used here were substantially the same as those previously described, but the following reference to what is probably one of the earliest laboratories run by an industrial concern may be of interest:

"A laboratory has been fitted up where they experiment on the different methods of converting copper into brass. It contains a large number of assaying ovens and furnaces and the machinery works by flowing water. There is a small hammer used for testing how many blows the brass can endure without breaking; there are also teeth with which the brass is struck, but only once at any one point."

Apparently at this time considerable interest was shown in other parts of the world in the possibility of replacing stone moulds with iron. At Ockran in the Tyrol, again according to Swedenborg, "the brass is not poured out between masses of stone as elsewhere, but upon an iron plate overlaid with watery clay and indeed into 31 thin plates or bars each weighing $4\frac{1}{2}$ lbs."; while, describing methods used at Hamburg, he states, "an

attempt has also been made to use iron instead of stones, but the result has been to make brass of the coarsest quality."

A few years later (1764), Galon ¹³ gives a description of brass ingot manufacture in Namur which, although essentially similar to that described by Swedenborg, is interesting for its illustrations. Figs. 2 and 3 taken from this paper show the interior of the foundry. Melting furnaces carrying eight or nine crucibles of calamine-copper mixture are shown at A, B and C. When the charge was ready the crucibles were removed from the furnace, skimmed in the pits G and H and transferred to the larger crucible for pouring into the mould. As soon as the pouring was complete the mould was hoisted to a horizontal position, the upper part lifted in the hoist end and the ingot removed. Another illustration in the original paper depicting the tools used in the foundry shows tongs, etc., almost identical with those used in present-day practice. Details of the mould construction given by Galon indicate that the material used at Namur was a "sort of granite" obtained from Mont St. Michel. Two flat slabs of this granite formed the main faces of the mould, three iron bars being placed between them to form the sides and bottom. Thus the size of the ingot and its thickness were governed by the size of the iron bars used.

It will be seen from these descriptions that the methods of brass strip ingot casting several hundred years ago were essentially similar to modern pit furnace practice. Improvements have been made in minor details, particularly in the refractories used for the crucibles and by the introduction of cast iron as the mould material. No information is available as to the date of the introduction of cast-iron moulds. It is known (Percy ¹⁴) that this was prior to 1860, and in all probability was much before this. The universal adoption of cast iron could only have been possible on the development of mould coatings more suitable than the "watery clay" referred to by Swedenborg.

No reliable information as to the later development of casting practice in detail in the nineteenth century has been found, probably because of the spread of the industry into many separate districts and the secrecy observed by the skilled workers and their employers regarding methods.

THE HISTORY OF BRASS

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CHAPTER II

CRUCIBLE FURNACE PRACTICE IN 1921 AND SUBSEQUENT DEVELOPMENTS

Moulds—Mould Coatings—Melting and Casting Practice—Electric Furnaces—New Types of Mould.

THE improvements effected in the casting process since the time of Galon could perhaps best be appreciated from a description of English crucible furnace practice at the time of the commencement of the researches described in this book. A careful survey of such practice in a number of different works showed considerable variation in detail, but the following description can be taken as representative of average English practice about the year 1921.

The moulds used were of grey cast iron of which the actual compositions varied considerably. Hæmatite iron was generally preferred, but analyses taken from a number of commercial moulds showed silicon contents varying from 1.5 to 4 per cent. with graphite from 2.5 to 3 per cent. Sulphur and phosphorus were generally low, but in one case phosphorus as high as 1.2 per cent. was found.

While the ingot varied considerably in width (from 3 to 14 inches) and length (from 24 to 46 inches), the thickness was commonly in the neighbourhood of 1 inch, the minimum being about 0.9 inch and the maximum about 1.5 inches. In nearly all cases some "camber" was given to the flat faces of the ingot mould in order to make the ingot surfaces slightly convex; the thickness of the centre of the ingot exceeded that at the edges by amounts up to 0.1 inch. The design and dimensions of a typical commercial mould are shown in Fig. 4. In this case the walls are of equal thickness, *i.e.* 1.5 inches, but in some works moulds were used in which the lid or top face was thinner than the back. The two halves of the mould were held together by rings and wedges or screw clamps. In general, the moulds

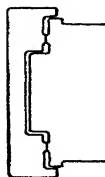
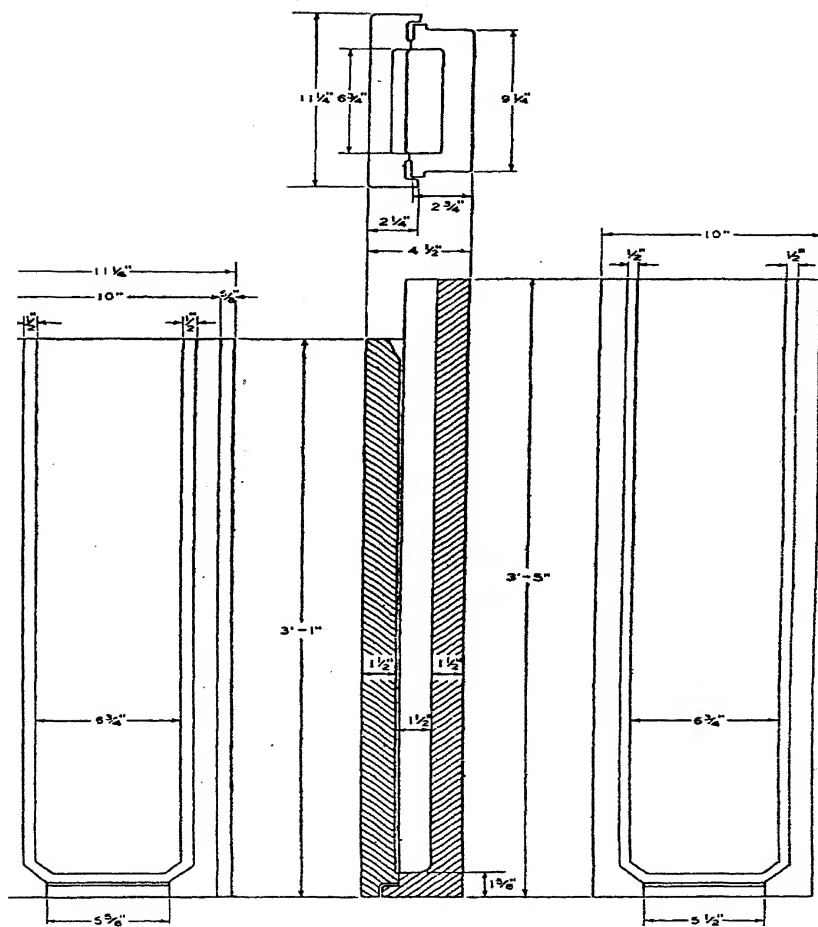


FIG. 4.
Dimensions of typical commercial strip ingot mould.

were used as cast, without treatment of the inner faces, but in some works the inner surfaces were machined.

When in use the mould was leaned against a casting bench at an angle of 15° to 30° to the vertical and the stream of metal necessarily impinged on the mould faces during pouring. Cracking of the inner surface layers occurred invariably after a few casts, and gradually extended, being the chief factor determining the life of the mould. The cracks were usually filled by clay or mould dressing before each cast to avoid holding of the ingot face and the consequent formation of contraction cracks or "pulls" in the ingot surface. It was general experience that the back or under-side of the mould, which was the more overheated locally during casting, deteriorated at a much quicker rate than the lid. The inner faces of the mould were generally prepared before casting by the application of a thick coating or "dressing" composed mainly of volatile and combustible material. The nature of the dressing used varied considerably, the commonest being oil, oil and powdered charcoal, oil and resin, molten resin, tallow and oil.

The effect of the mould dressing is to produce on contact with the molten metal a large volume of gas and flame which envelops the stream of metal and the mouth of the crucible, thus protecting the metal stream from contact with air. In many cases extra quantities of dressing were applied to the bottom of the mould in order to increase the initial evolution of gas.

Melting was carried out in large single crucibles, generally in coke-fired pit furnaces, the materials used being virgin copper and zinc with varying proportions of process scrap. Gas- or oil-fired furnaces were occasionally used. In making up the total crucible charge, which varied from 140 to 200 lbs. in weight, scrap and copper were first melted together, and when the charge was completely molten and raised to a sufficiently high temperature the necessary amount of spelter was added. Fluxes, where used, varied considerably in composition, but salt was the most generally favoured material. This was added to the molten charge either before or after "speltering" or immediately before pouring, but in some cases charcoal, with or without salt, borax or other flux, was added with the crucible charge.

When the pouring temperature was reached the crucible was withdrawn from the furnace by means of pulley tackle and taken to the casting bench against which the moulds were leaned. The metal was skimmed to remove flux and dross from the surface and the pot then tilted by the caster standing in front of the mould, the edge of the crucible resting on the rear edge of the mouth of the mould. This method of casting is illustrated in Fig. 5, where the large flame from the burning mould coating is clearly shown.

In general the temperature of casting was as high as could conveniently be reached without undue loss of zinc, though the maximum was not always obtained. The actual temperatures for different compositions were approximately as follows: 70/30—1100° C.; 65/35—1070° C.; 62/38—1050° C. The rate of pouring varied in different works and with different sizes of ingot; the larger the mould the greater was the speed of pouring generally used. It was found, however, that when measured as the average rate of rise of the metal surface in the mould the speeds were more nearly constant and in all cases came within the limits of 1-2 inches per second. Usually the speed varied during the pouring of any one ingot, the bottom portion of the mould being filled at a greater rate than the upper part.

Narrow ingots were poured in a single stream, but where the width of mould exceeded 4 or 5 inches the stream of metal was divided by means of an iron rod or plug of carbon held against the crucible lip. Solidification shrinkage (piping) was fed by small additions from the crucible and in no case was a hot dozzle or feeder head employed. Shortly after casting the mould was loosened, the ingot removed and allowed to cool in air.

At the present time the general features of crucible brass casting are similar to those just described, though progress has been made by the introduction of improvements in a number of works. A large output of brass strip is now made, however, from ingots cast from electric furnaces,* and the engineering

* Although most of the main types of electric metallurgical furnace are of European origin and their use in manufacture has made great progress in several European countries,¹ the development of the electric melting furnace and its application to commercial practice was most rapid in America, where (according to Gillett²) the first furnace was

and economic factors connected with large melting units have necessitated modifications of the crucible method of casting, some of which are important from the metallurgical point of view.

An essential feature of electric plants designed for high efficiency and continuous production is the tilting of the furnace, and the moulds must be brought below the spout of the actual furnace for efficient working. When casting strip ingots of the usual size this is achieved by mounting the moulds round a turn-table. An intermediate tundish or pouring bowl is necessary to direct the stream into the mould, and this in turn requires the mould to be fixed vertically.

After pouring one ingot the table is rotated to bring the next mould into position. Cast-iron moulds with hinged lids, each with a hinged tundish, are used, the inner surface being coated with a volatile dressing as in crucible practice. The main differences, therefore, between the electric furnace and crucible practice, apart from uniformity in temperature, size of stream and rate of pouring, are that the moulds used with the electric furnace are stood vertically and that the stream of metal is distributed over the width of the ingot by means of a tundish.

In recent years important modifications have also been made in ingot moulds. In 1926 Junker⁴ introduced in Germany copper-faced water-cooled moulds. These will be discussed in detail in a later chapter, but they are significant in connection with casting methods in that such moulds can conveniently handle very large ingots and are being used to take the whole of the charge of an Ajax-Wyatt furnace to give one

installed in 1906 for steel production. The utilisation of electric furnaces for brass making did not begin until about 1916, but from that date expansion in this direction was rapid and Gillett claimed that in 1927, 625 electric furnaces were in active use in the production of brass and that these were responsible for 90 per cent. of American brass ingot production.³

For brass rolling mill practice working on continuous operation the vertical ring induction furnace of the Ajax-Wyatt type is now almost universally used. In this country the development of electric furnaces for melting was slower than in America, and in 1921, when the detailed survey of British works practice in brass production was made, the only electric furnace in use for the purpose was a Baily (resistance type) furnace. The first Ajax-Wyatt was installed in 1922, and at the present time there are 74 such furnaces in Great Britain. The majority of these are of 600 lbs. capacity, rated at 60 kw., but three have been installed of 25 cwt. capacity.

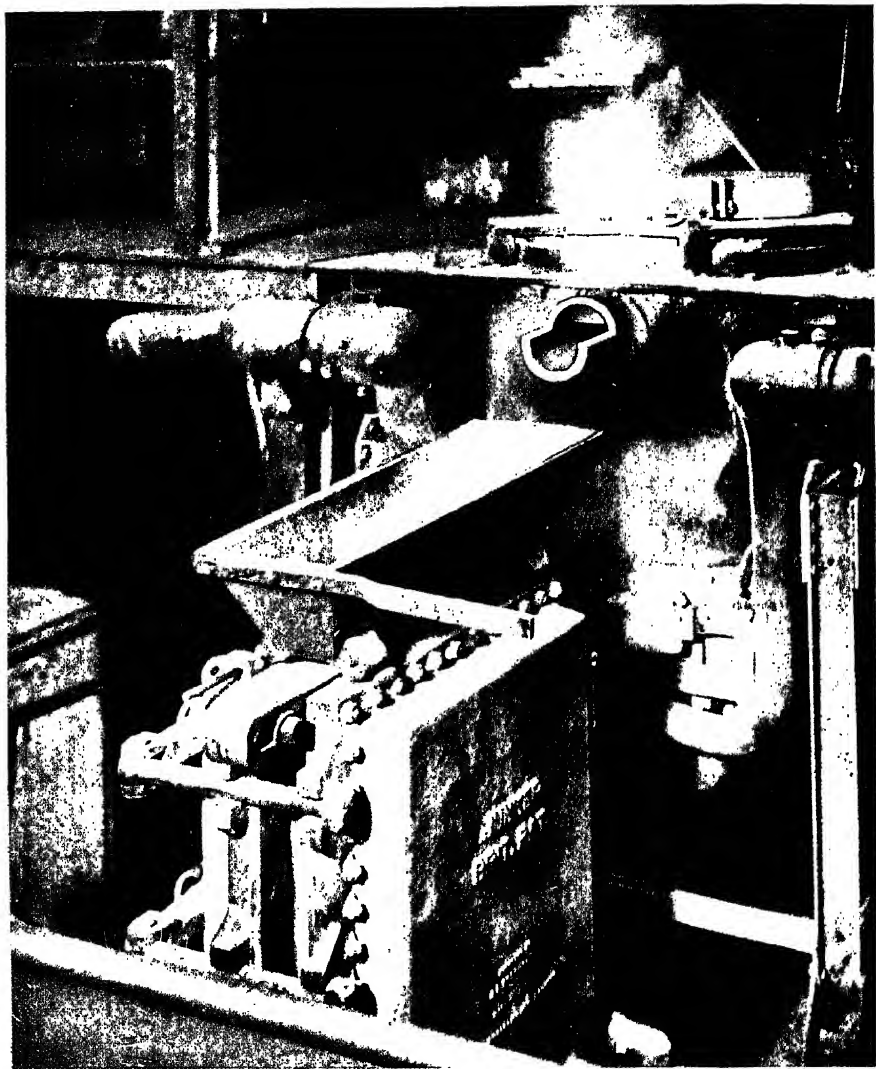
PLATE IV.



[Courtesy of the Carborundum Co., Ltd.]

FIG. 5.

Pouring of brass ingot in crucible practice.



[Courtesy of Messrs. I. C. I. Metals, Ltd.]

FIG. 6.

Ajax Wyatt electric furnace, capacity 1000 lbs. brass, with Junker copper-faced water-cooled mould of similar capacity.

slab ingot of up to 1000 lbs. in weight. The use of such large ingots depends obviously on other factors, such as facilities for hot rolling, but the casting of a single ingot instead of a number of smaller ingots results in a great saving in pouring time. The method of casting into these large moulds of the Junker type is essentially the same as that employed with the smaller cast-iron moulds on the turntable and volatile mould coatings are used. A modern English plant showing a 1000 lbs. capacity Ajax-Wyatt furnace and a copper-faced water-cooled mould of similar capacity is illustrated in Fig. 6.

More recently a new type of water-cooled mould has been advocated by Erichsen. This differs from the Junker mould in that the working faces are of a low conductivity nickel-iron alloy and are mounted in such a way that their distortion after casting compresses the ingot and avoids any tendency to piping. These developments of mould practice and their metallurgical aspect are discussed fully in a later chapter.

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CHAPTER III

BRASS STRIP INGOTS.* CHARACTERISTICS AND DEFECTS

Ingot Dimensions—Surface—Structure—Internal Soundness—Micro-structure—Non-metallic Inclusions.

IN the production of any type of casting the actual shape required is a factor of considerable importance, and as a general rule, masses of large section and short length present least difficulty in the casting of internally sound and externally smooth ingots. It has long been realised that the slab form of ingot used in brass manufacture (commonly having a thickness of only about 1 inch) is unfavourable to the production of material of the highest quality. The reason for the adoption of the slab form was mainly the resultant saving in rolling coupled with the fact that all rolling of 70/30 brass was done in the cold. Subservience to these economic factors was unavoidable and it is remarkable that in practice brass foundrymen have been and are still able, by accommodating their methods, to produce in ingots of this unsuitable form, brass of a high average standard of quality. The chief disadvantage of the thin parallel slab ingot is the rapidity of solidification of the metal through the ingot section, which precludes to a large extent any measures for control of shrinkage or for facilitating the escape of gases and included matter. This was one of the points brought out by some early work on the subject at Woolwich,¹ in which a practical trial was made to compare the qualities of vent-sealing tubes made from ingots of different cross-sections; it was found that defects consisting of non-metallic inclusions which caused a high proportion of failures in tubes made from long thin ingots were entirely avoided by the use of much thicker and shorter

* The term "Strip Ingot" is used to describe an ingot from which strip or sheet is rolled.

ingots. A further disadvantage of the slab ingot is the relatively large area of surface which must be cast free from folds and other defects.

These features render the problems of strip manufacture definitely more difficult than those met with in the production of bars. A method of producing sound strip while retaining the advantages of the slab form for rolling is understood to have been tried some years ago, using slices cut from ingots of large square section. The method was apparently not commercially possible and no case of its regular application is known.

Although as a result of the introduction of hot-rolling methods (see p. 13) the small form of slab ingot has been superseded in a section of the industry by a thicker and heavier ingot of similar proportions, it may still be regarded as common practice. On the whole a definite improvement in ingot quality would be expected from the larger ingot, but the disadvantages of the slab form are not wholly eliminated, and the controlling factors remain substantially similar in the two types. There is also general evidence that even when the long thin form of ingot has been retained, the quality has been improved where electrical melting in large capacity furnaces has been adopted, probably owing to the modifications in casting technique involved.

The work of the skilled practical caster is largely judged by the quality of the rolled or further manufactured product, and in the light of long experience he can diagnose certain types of defect connected with foundry manipulation. In many cases, however, lack of knowledge as to the true character of faults in the manufactured material leads merely to more or less blind changes in details of practice, and frequently the occurrence of a batch of strip of poor quality has been erroneously connected with some coincident change in the source of raw metals.

In the course of their investigations on brass the authors have had the privilege of obtaining casting data and examining in detail ingots cast under recorded conditions from the current output of a large number of works. A survey of this kind is valuable in affording an impression of the variations in quality which exist, and their connection with practice, and in

suggesting the most suitable lines for systematic investigatory work.

Commercial ingots made by only slightly different methods of casting vary considerably in surface quality. The principal causes of the surface irregularities, judged from visual indications, appear to be the use of worn moulds and the retention by the ingot surface of small shots of metal produced by turbulent pouring. In general the surface quality of ingots produced by the ordinary crucible casting process is better in the lower half of the ingot, while the front or lid surface is smoother than the back. These general observations may be explained by the great rapidity of pouring during the early stages of casting, by the fact that the stream of metal impinges on the back of the mould, producing overheating and local wear, and by the effect of inclination of the mould in giving folds more readily at the lower back face where the angle of incidence of the metal surface and the mould face is relatively acute.

At a position coinciding with the impingement of the stream on the mould face, an ingot defect may occur consisting of surface holes (or in an incipient form as local roughness), known as "blowing," which may be sufficiently pronounced to produce folds in rolled strip. The first ingots cast in moulds which have been rested for a period (even for only a few days) invariably show a rough surface (incipiently blown), but the second and succeeding casts are of normal quality. It has further been observed that certain moulds give blown ingots more consistently than others, and foundrymen have rightly attributed the trouble to some obscure effect occurring at the mould surface. The study of this feature of cast-iron moulds and the mechanism by which blowing occurs in brass ingots are described in Chapter XII dealing with mould materials.

Consideration of the quality of strip ingots, together with an analysis of the relevant casting data, affords evidence that a good surface is producible only by the use of a high casting temperature and a total pouring speed not much lower than 1.5 inches of ingot length per second. Where a low temperature is used the ingots are generally affected either by folds or by the inclusion in the surface of flattened globules of metal which were originally splashed on the mould face and subse-

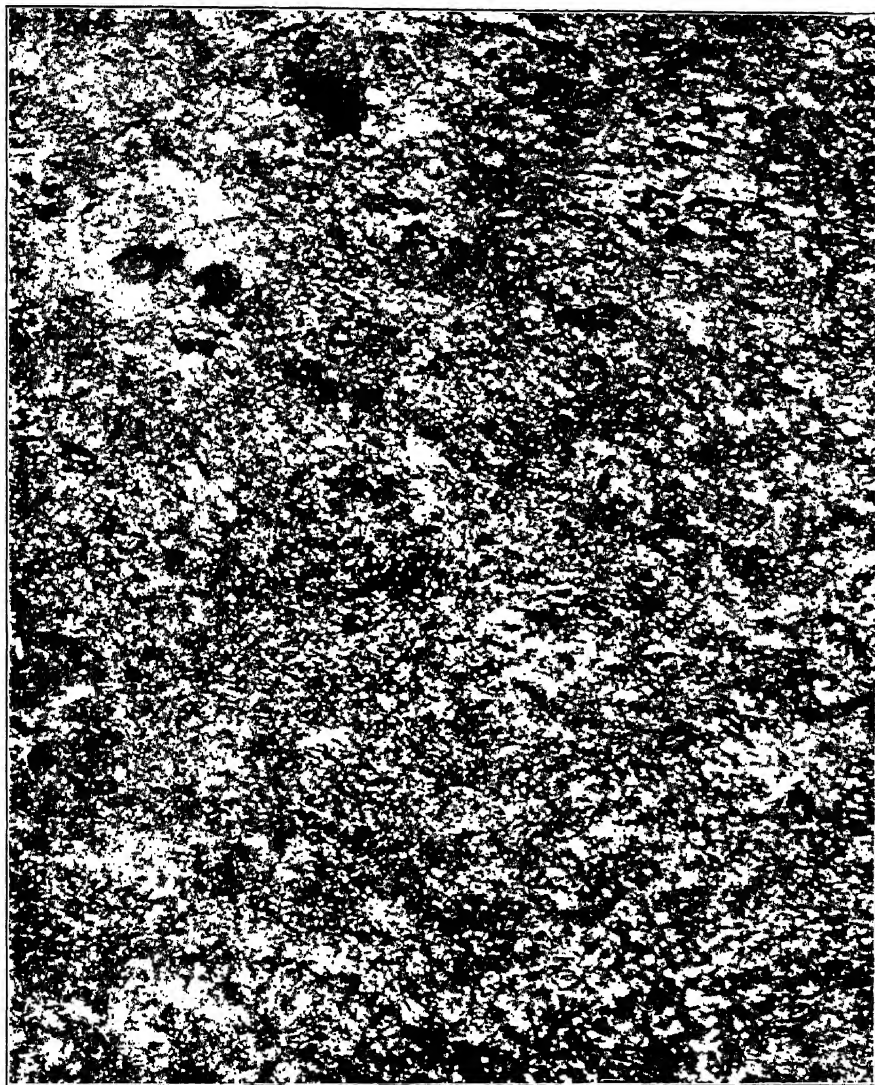
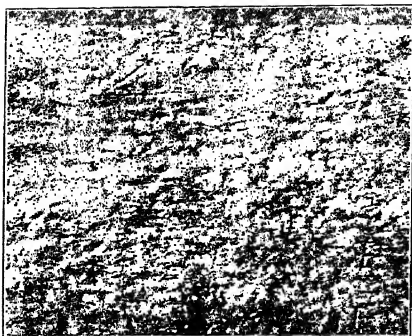


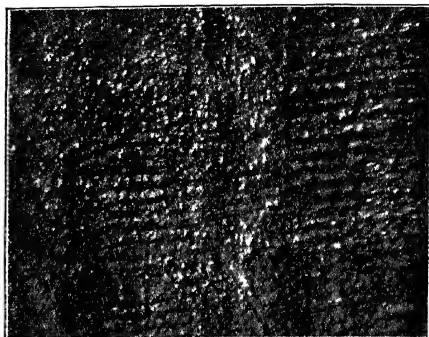
FIG. 7.

Surface of brass ingot cast at low temperature. (Ingot H, Fig. 8.) $\times 2$.

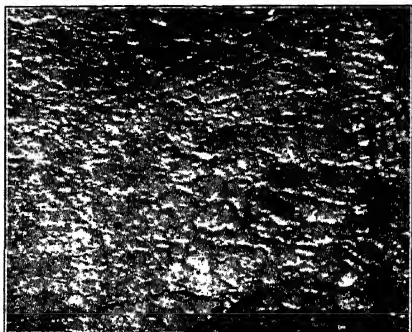
PLATE VII.



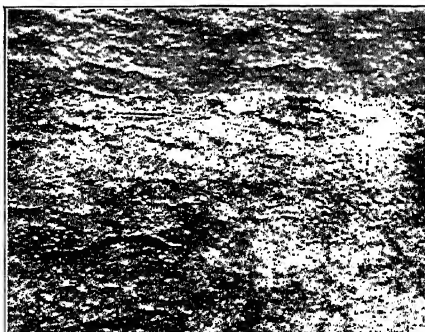
Ingot B.



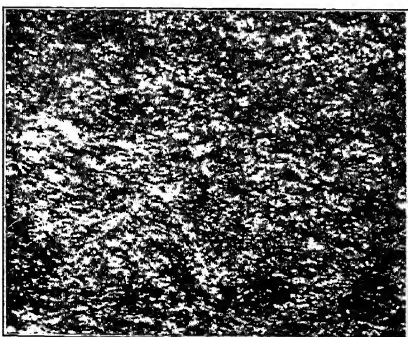
Ingot C.



Ingot D.



Ingot E.



Ingot H.



Ingot J.

FIG. 8.

Surfaces of representative brass ingots showing variations produced in commercial

quently partially remelted into the mass of the ingot (see Fig. 7). The photographs, Fig. 8, show the surfaces of a number of brass ingots representative of the ordinary variations produced in industrial practice. Details of the casting conditions and compositions are given in Table I below. Further reference is made to the same ingots later (p. 20), in connection with the distribution of unsoundness.

TABLE I
COMPOSITION AND CONDITIONS OF CASTING OF TYPICAL
COMMERCIAL BRASS INGOTS

Ingot.	Ingot size, inches.	Copper content, %.	Typical Casting Conditions at Supplier's Works.	
			Approx. Casting Temp. °C.	Pouring speed. Rate of rise of metal in mould in inches/sec.
A	35 × 9 × 1.25	65.2	1080	1.5-2
B	31 × 7.5 × 1	69.4	1100	2
C	34 × 8 × 1	64.6	—	1.4
D	39 × 5 × 1.1	62.0	1050	1.6
E	31 × 9.5 × .88	70.9	—	1.2
F	27 × 7 × 1	68.8	—	1.3
G	22 × 12.7 × .75	62.4	—	1.4
H	28 × 8.5 × 1	68.1	1070	1.2
JE	31 × 11.5 × 1.6	62.2	—	1.3
(Electric Furnace)				
JC	„	62.6	—	—
(Crucible)				

The macrostructure of 70/30 brass ingots, revealed by etching longitudinal and transverse sections, is in all cases variable, and few well-defined regions of columnar or of equiaxial free crystals can be distinguished. This variation in structure is shown later to be connected with the movement of the metal in the mould during casting. Generally the extreme outer layer of the ingot consists of very fine equiaxial crystals enclosing a zone of irregular columnar crystals, with a further central region of both coarse and fine equiaxial crystals. The bottom portion of the ingot tends to be more columnar in structure than the upper portion, which is frequently almost

entirely equiaxial although markedly irregular as regards crystal size.

Ingot sections showing the variations commonly met with are illustrated in Fig. 9. It is difficult to regard any illustration of a single ingot as typical, since extremely wide variations may be found in commercial ingots cast by methods which apparently vary only in small detail. Further, there appears to be no relation between the structure of the ingot and comparatively wide variations in practice such as occur when both crucible and electrically melted brass are cast in the same works.

A definite connection does exist, however, between structure and composition. With brasses of low copper content (*i.e.* about 62 per cent.) the structure is more columnar in type, and in very thin ingots of this composition, the entire structure has been found to consist of columnar crystals with the exception of the extreme thin outer layer of fine "chill" crystals, and a portion at the top showing the normal variable structure (see Fig. 10).

As regards internal unsoundness, brass strip ingots, although differing widely, invariably show three types of cavity, distinguished as contraction cavities, subsurface cavities, and minute distributed cavities of similar form.

Contraction cavities of characteristic shape are shown in Fig. 11. They are recognised by their irregular and frequently jagged outline, due to the fact that they consist of gaps left between the dendritic (tree form) crystals of brass. These are produced near the end of solidification of the ingot mass by the sinking away of the intercrystalline liquid to feed the contraction occurring in neighbouring regions. They occur always in the central portion of the ingot and may often be detected in longitudinal sections over almost its whole length. Their character varies considerably according to the composition of the brass. In brass of 70 per cent. copper content the cavities cover a central band of appreciable width but are small and are commonly referred to as "sponginess." As the copper content is decreased the band of sponginess tends to diminish in width. At a content of about 62 per cent. copper the contraction cavities are concentrated on the central plane of the ingot marking the meeting-point of the columnar crystals

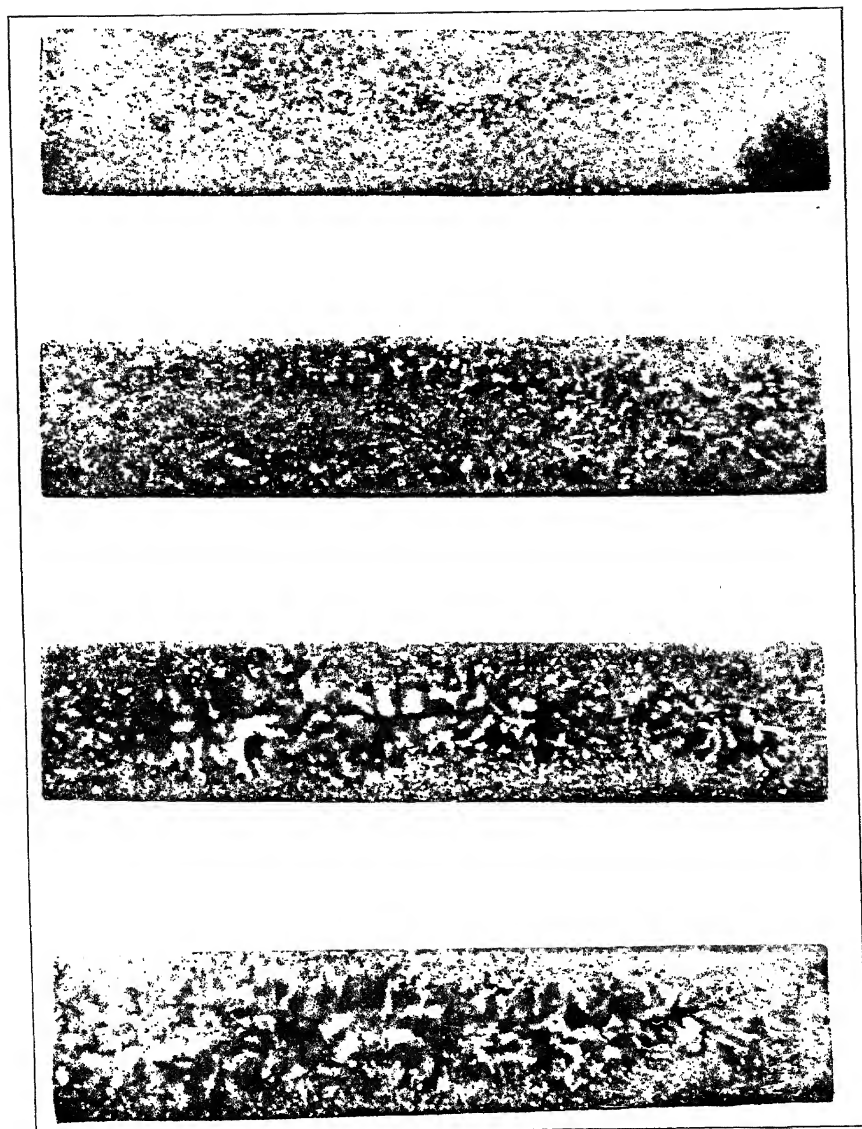


FIG. 9.

Transverse half-sections of ingot B (69% Cu) showing typical variations in structure along the length. Actual size.

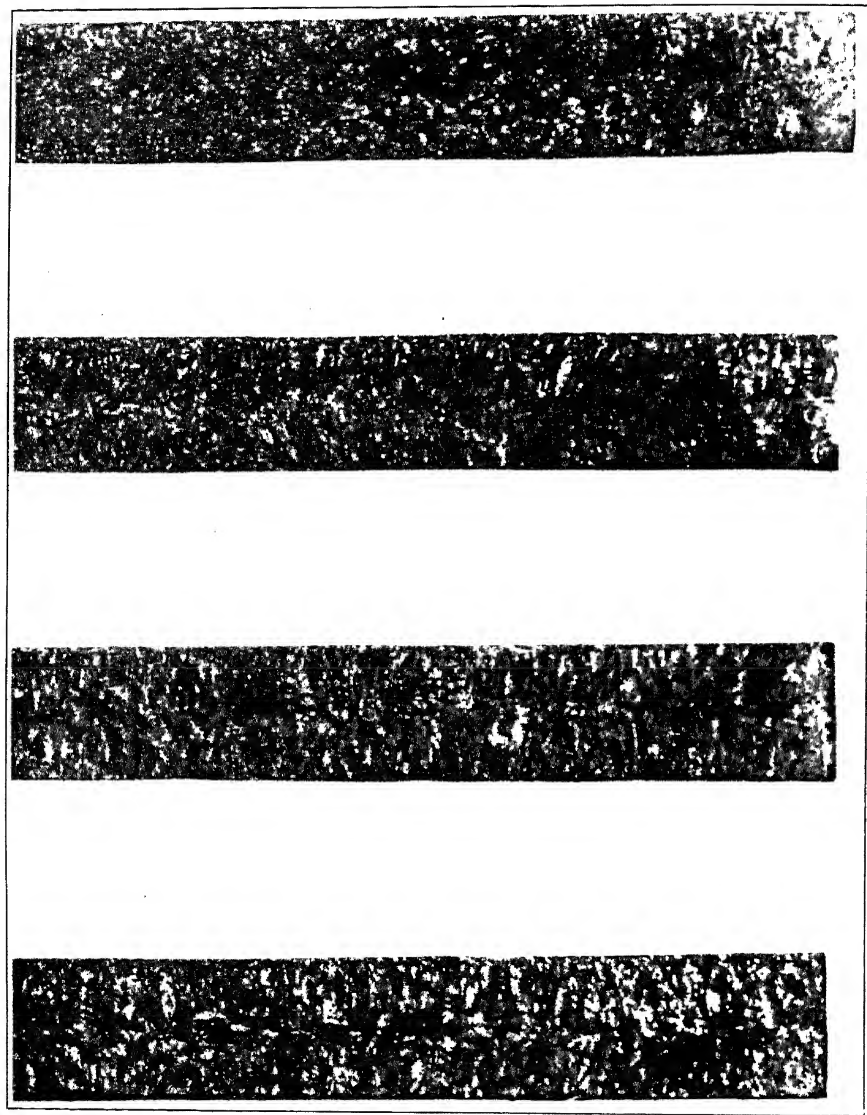


FIG. 10.
Transverse half-sections of ingot G (62% Cu) showing columnar structure.
Actual size.

PLATE X.

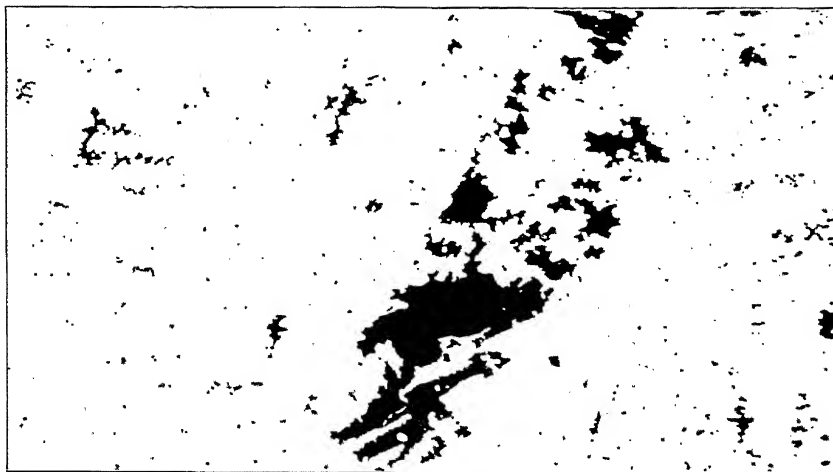


FIG. 11.
Contraction cavities in 70/30 brass ingot showing characteristic form.
× 25.

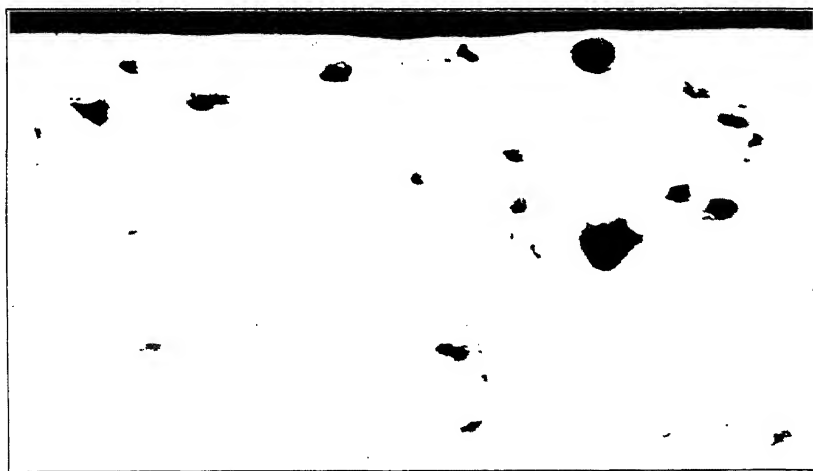
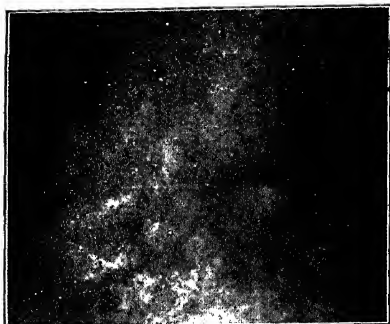
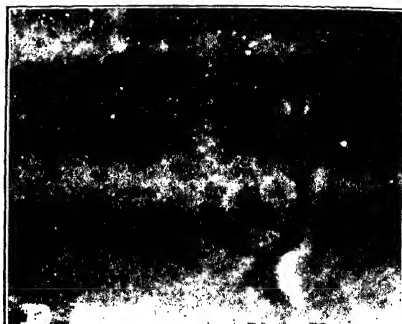


FIG. 12.
Subsurface cavities of entrapped gas. × 25.

PLATE XI.



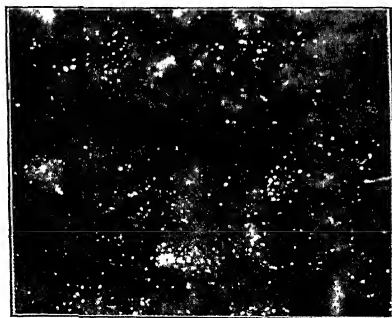
Ingot B.



Ingot C.



Ingot D.



Ingot F.



Ingot G.

Fig. 13.
Radiographs of surface layers of representative brass ingots corresponding to
showing surface irregularities and subsurface unsoundness. Actual size.

growing from the two faces. In general, shrinkage cavities are most marked in the lower half of the ingot; this is not unexpected in view of the difficulty of feeding contraction over a narrowly constricted long length, and of the normal crucible casting practice of filling the bottom portion of the mould at a greater rate than the upper portion in order to secure good surface quality at the lower end of the ingot. Examination of various types of contraction cavity shows some to be internally clean while others are discoloured.

The form of cavity referred to by the term "subsurface cavity" is always roughly spherical in shape. Such defects occur distributed throughout the ingot but are mainly found near the surface. Typical examples are shown in photomicrograph Fig. 12. The shape and mode of distribution of such cavities suggest that they were formed by gas either entrapped by, or evolved from, the metal. It is clear from the number and size of the subsurface cavities which may occur in strip ingots that surface defects on rolled strip arise largely from this source.

The use of X-rays for the examination of whole ingots, although possible, is not capable of revealing with distinctness individual small cavities such as are being described. For investigatory purposes, therefore, the method has been adopted of examining by X-rays thin sections of ingots cut parallel to the surface, one face of the section being the original ingot surface. This is the most satisfactory method available for determining the subsurface quality of ingots and for illustrating the degree of irregularity of the ingot surface. The X-ray opacity of a metal increases in proportion to the thickness, so that by taking specimens of such dimensions that the depth of the surface irregularities is appreciable compared with the thickness (0.125 inch is suitable), the irregularities are outlined in the radiograph and their relative depth is shown by the amount of contrast produced in the radiograph.

Radiographs of the specimens from ingots previously illustrated in Fig. 8 are shown in Fig. 13. The smoothness of ingot B is shown clearly by the flatness of tone. In ingots D and H deep depressions are shown by the lighter areas. Small white circles (in ingots D, H and J) indicate the presence of subsurface cavities and illustrate the variations in size and number

which may occur in a relatively small portion of the ingot. The size of cavity which can be detected with certainty in a layer of metal 0.125 inch thick cut from an ingot surface has not been determined accurately, but in view of the liability of surface roughness to obscure the underlying unsoundness, it is estimated that only internal cavities of more than 0.02 inch diameter are clearly visible.

To detect very small cavities it is necessary to prepare specially machined surfaces for visual examination. The method used by the authors consists in machining the ingot surface parallel to the face with light cuts of about 0.01 inch, using a very sharp tool which will not burr over any cavities present. This provides a means of examination which, although laborious, is more critical and informative than that of cutting transverse sections. A typical section prepared in this way is illustrated in Fig. 14. Transverse sections cannot expose a sufficient number of internal cavities to give any quantitative impression of the degree of unsoundness present.

The examination of large numbers of commercial ingots by these methods has shown considerable variation to exist in the number of subsurface and other cavities of similar form distributed through the ingot thickness. In general, the lower the copper content the sounder is the ingot, while the use of a high casting temperature tends to promote freedom from subsurface defects. Small cavities of 0.01 inch or less in diameter are apparently present in appreciable numbers in all top-cast brass ingots, and, unlike the larger type, are not confined to the surface layer, but are distributed fairly uniformly through the ingot thickness.

In order to map out quantitatively to some extent the variation in the degree of unsoundness throughout the ingot the method of density determinations on cylinders cut from various positions has been adopted. Fig. 15 gives the results of density determinations on a number of ingots from various sources (see Table I), the surface and subsurface characteristics of which have been previously illustrated in Figs. 8 and 13. These results in conjunction with a knowledge of the method of manufacture show certain features of general interest. The extreme bottom portion of each ingot was sound and with one exception the metal at the ingot edges was also nearly of

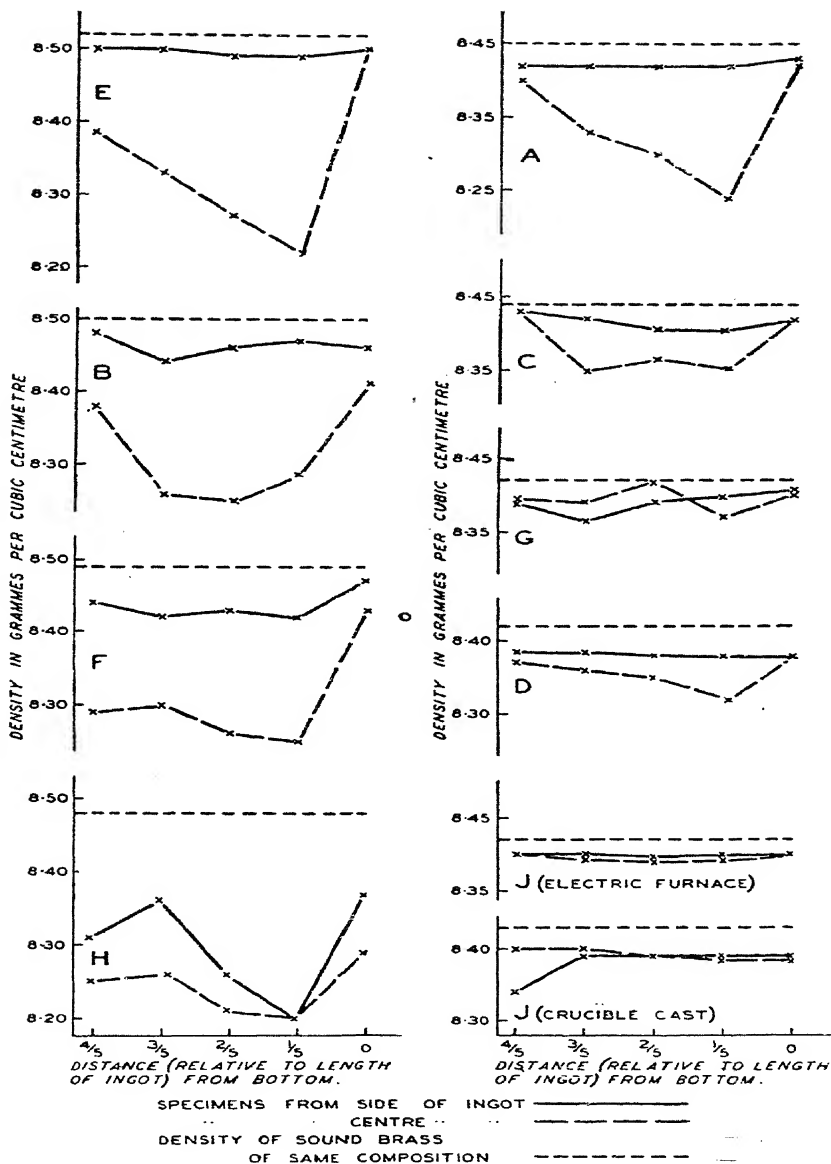


FIG. 15.

Density variations in typical commercial ingots.

the maximum density of the alloy. Maximum unsoundness was in general found just below the middle of the ingot, while the top regions were relatively sound. It is clear that the absence of defects from a fracture made by breaking off the head of the ingot, although showing that the "pipe" has been fed, does not give any indication of the quality throughout the remaining portion.

A further feature of the density values is the high degree of soundness shown by ingots of relatively low copper content. This, together with the visual indications, suggests that in brass of this composition the unsoundness is not only almost entirely concentrated on the central plane but is less in total amount than in ingots of higher copper content in which the contraction cavities appear as a region of central sponginess. Further reference is made to this effect in Chapter VII.

The microstructure of brass ingots requires little comment here additional to the matter given in Appendix B. Strip ingots show without exception a microstructure normal for chilled castings, the only feature of note being the presence of the beta constituent, due to the rapid rate of cooling, in all ingots containing less than 70 per cent. of copper. The cavities constituting central unsoundness are definitely of interdendritic type, while those occurring near the outer surface are entirely independent of the crystal structure.

Commercial ingots show a considerable variation in the quantity of included non-metallic matter, but without any definite relation to the known variations in casting practice. If the ingots A to J already described may be regarded as typical for the purpose, it would appear that brass of only 62 to 63 per cent. copper, whether crucible or electric furnace cast, contains fewer inclusions than the alloys of higher copper content. Otherwise the inclusions appear to have no connection with the mode of solidification of the ingots and their distribution follows no uniform plan. The colour of the majority of inclusions found in brass is dark grey, not dissimilar from but somewhat lighter than that of lead. The quantity of lead present is in general, however, much less than would account for the number of inclusions observed, and the only possible materials composing them are zinc oxide and slaggy non-metallic material.

PLATE XII.

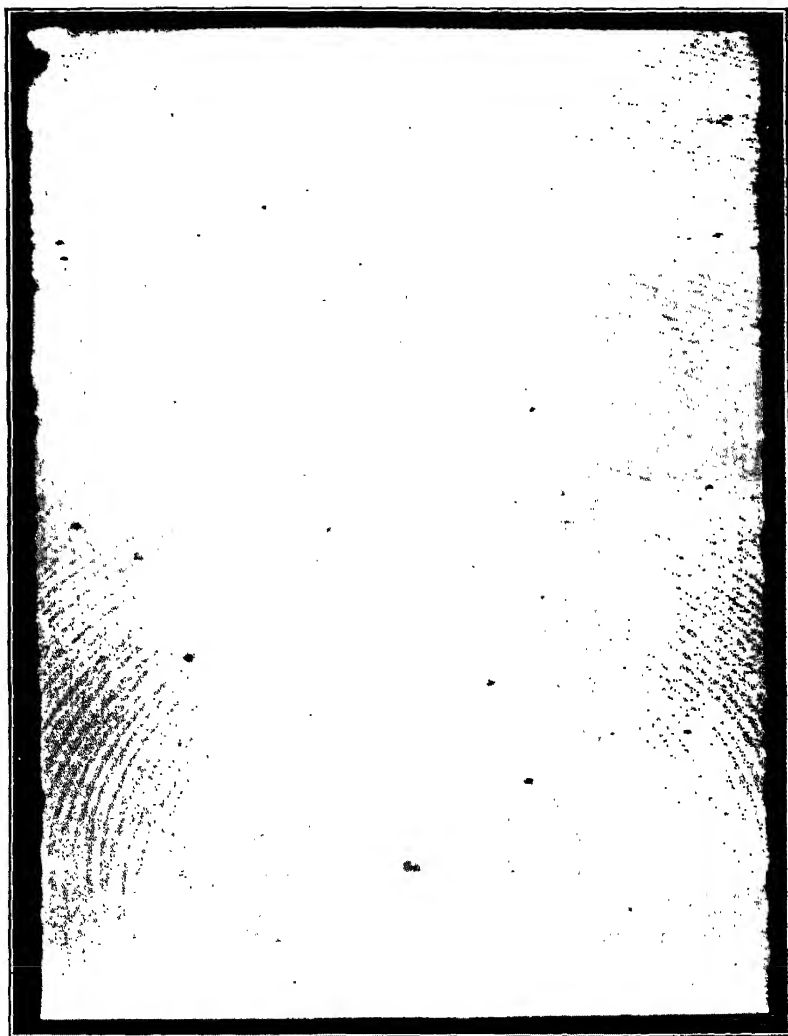
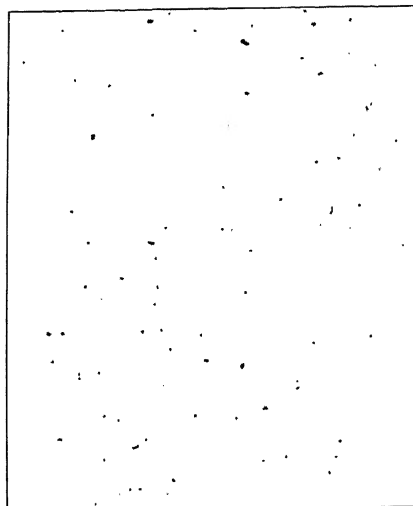
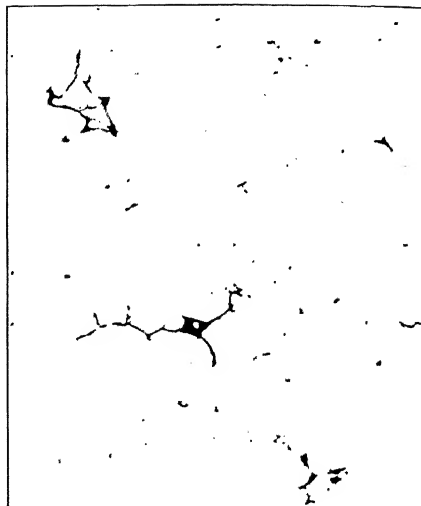


FIG. 14.

Section illustrating method of fine machining to reveal minute cavities in ingots.



(a) $\times 100$.



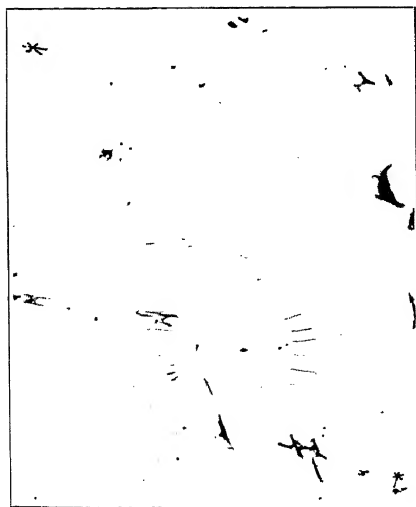
(b) $\times 100$.

FIG. 16.

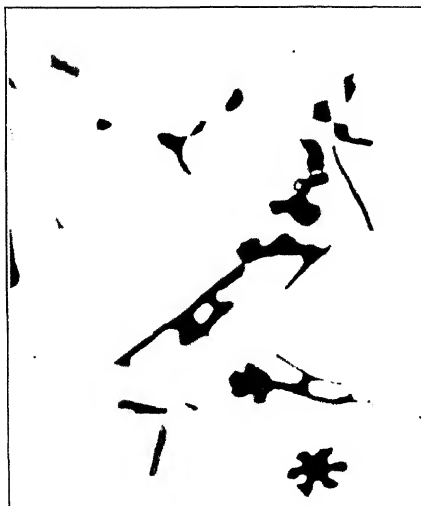
Typical non-metallic inclusions in brass ingot.

(a) Ingot 62% Cu.

(b) Ingot 68% Cu.



$\times 100$.



500.

FIG. 17.

Synthetic zinc oxide inclusions in 70/30 brass.

Zinc oxide itself when introduced into molten brass is entirely unwetted by the metal, probably due to the effect of adsorbed air, and it has been found impossible to prepare artificial zinc oxide inclusions except by the addition of cuprous oxide to react with the zinc in the brass and give zinc oxide in contact with, and wetted by, the molten alloy.

When cuprous oxide is thus introduced into a small melt of 70/30 brass, and the metal allowed to solidify rapidly, micro-sections from the resulting mass show numerous inclusions identical in characteristics with those observed in ingot sections. The inclusions remain *in situ* during the polishing operations, showing no tendency to dragging of the surface, whereas similarly produced inclusions of borax and glass give much more difficulty in polishing, leaving small holes and scratches in the surface. This evidence would suggest strongly that the inclusions found in ingots are mainly composed of zinc oxide.

In other experiments designed to give information as to the degree to which inclusions of zinc oxide may normally remain in molten brass, melts were made in a long vertical cylindrical furnace of small diameter, cuprous oxide being introduced at the bottom of the melt. The brass was maintained molten for varying periods before solidification. It was shown that zinc oxide inclusions produced in this manner rose with great rapidity to the top of the melt and that in a mass of molten brass at rest no zinc oxide inclusions of appreciable size can remain suspended except momentarily. The conclusion is, therefore, that the actual inclusions found in brass ingots have been formed mainly during the pouring of the metal and would be likely to be eliminated completely by a small retardation in the rate of solidification.

Typical inclusions in brass ingots are shown in Fig. 16. Synthetic zinc oxide inclusions prepared in a small melt and allowed to solidify in the crucible are shown in Fig. 17. These are formed freely in the molten brass and are consequently different in shape from those formed in ingots and subjected to turbulence during casting.

The chemical estimation of zinc oxide in brass by reduction with hydrogen is complicated by the volatility of zinc at the high temperature required and a modified method was devised

in connection with the authors' work by Evans and Richards.² In this method hydrogen is passed over the heated brass in a suitable apparatus and the volatilised metallic zinc is collected on copper gauze, the water vapour resulting from reaction of the zinc oxide with hydrogen being suitably collected and weighed. Determinations by this method indicate that in commercial brass the content of zinc oxide does not normally exceed 0.025 per cent.

This amount is very small and suggests that entire freedom from inclusions could be obtained by only slight variation of normal casting practice in the direction of delaying the commencement of solidification. Non-metallic inclusions, however, seldom occur in sufficient quantity to be a source of defects in brass ingots, and in this respect the problems connected with brass casting are simpler than those of steel manufacture, in which the effects of inclusions on the mechanical properties of the material constitute a factor of great importance.

BIBLIOGRAPHY

¹ R. Genders, *J. Inst. Metals*, 1921, 26, 139.

² B. S. Evans and H. F. Richards, *J. Inst. Metals*, 1926, 35, 173.

CHAPTER IV

BRASS STRIP. SURFACE QUALITY AND TYPES OF DEFECT

Ingot Surface Defects—Spills—Blisters—Lines—Laminations—Methods of Testing for Internal Unsoundness—Correlation of Properties of Ingot with those of Rolled Strip.

SOME detailed reference to rolled strip is necessary at this stage as a supplement to the description of ingots, in order to follow the effects of rolling on the various features of the cast material and to assess the relative importance of different ingot defects in their ultimate form. Such considerations largely govern the lines on which improved methods are to be sought and the degree to which compromise is practicable in new methods.*

The first requirement of rolled brass is an immaculate surface, since the material is generally used in cold drawing or forming operations for the manufacture of articles required to have a good surface finish. Even the most critical inspection, however, may fail to disclose all defects, and it is clear from practical experience that in the subsequent working operations it is possible to develop hidden faults into blemishes which render difficult any final polishing of the finished article. While many of these may originate in invisible surface defects, certain of them undoubtedly arise from internal defects in the strip, and it is thus evident that any measures to improve the quality of rolled brass must be devoted to the control of both surface finish and internal soundness.

While inspection methods are variable in detail, according to the nature of the product and the works organisation, rolled brass is always examined at least at two definite stages. These

* The development of rolling mills and general methods employed in the manufacture of brass strip and sheet are described in Appendix A, p. 147.

are after the breaking down of the ingot (about 50 per cent. reduction in thickness) and after finishing. Whether the ingot is machined or not before or during rolling, a close visual examination after the first few passes is carried out to enable any surface defects which have developed to be chipped out. Such defects originate in flaws in the surface of the cast ingot, and if removed with due care form smooth hollows which later roll out perfectly, without any effect on the finished strip. Planing or scalping of the ingot as cast or after the initial passes avoids the bulk of work of this kind, and may prove more economical where heavy thick ingots are made, or where the production of a good ingot surface with a particular alloy is normally difficult. There is also much general evidence from experience, that strip with a superior surface finish is produced more regularly where scalping or planing is employed as a routine process.

Defects which occur in the early stages of rolling, being generally identified with pits, splashes or other irregularities in the surface of the cast ingot, are thus purely mechanical in character, representing in general the occasional defective spots liable to form locally in the otherwise smooth ingot surface. If completely removed at an early stage of rolling, their significance from the point of view of the finished product is negligible, but the possibility of a proportion of them persisting and the cost of removal are important considerations, and the total avoidance of such defects by the production of a faultless ingot surface would constitute a definite gain in economy of manufacture.

Inspection of the finished sheet or strip involves final rejection or acceptance. Defects present at this stage are, with some exceptions, developments of previously invisible faults which have been revealed by further rolling. The elimination of such internal defects consistent with smoothness of ingot surface can only be achieved by a complete understanding of the casting process.

Large quantities of rolled brass have been examined during the course of investigations on such material, and the visual indications have proved valuable as a preliminary to the detailed metallurgical examination of representative samples of good and defective strip from different works, and of ingots

PLATE XIV.

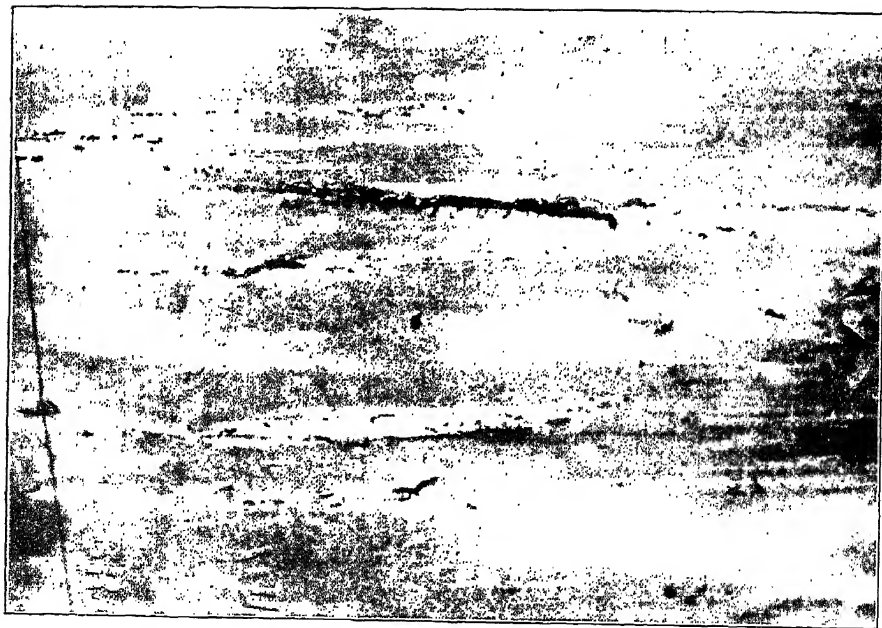


FIG. 18.
Typical "spill" defect in rolled brass strip. Actual size.

at various stages of rolling. The characteristics of the common types of defect are described below.

Spill.—This defect appears on the strip as an incompletely adhering film of metal, divided from the main surface of the strip by oxidised material and frequently capable of being peeled away, leaving a dark-coloured oxidised depression. Its name may be assumed to have originated from the fact that faults in the ingot surface, due to spilling or splashing of the metal during pouring, generally produce spills, and the same cause has long been assigned also to the small variety of the same defect commonly found in finished strip. Typical spill defects of pronounced form are shown in Fig. 18. Although quite small spill defects are as a rule easily detected on the smooth surface of the brass, certain types may be barely visible; these are invariably very narrow, though frequently extensive enough to cause rejections in subsequent manufacturing operations.

Owing to the oxide associated with spill defects, pickling in the usual sulphuric acid bath generally produces red stained areas about them, and the examination of strip in the pickled condition frequently reveals defects otherwise not readily visible on the rolled surface.

Few generalisations can be made regarding the occurrence of spills. The defects are frequently isolated or locally grouped with no exact reference to any particular position in the ingot. In most cases, however, strip or sheet made from ingots cast by the crucible method is markedly more defective in the portion corresponding to the front or "lid" face of the upper half of the ingot than elsewhere. Electrically melted and vertically cast material does not show such a consistent difference between the two surfaces. It is fairly general experience in manufacture that during the rolling of strip the number of defects tends to increase as the thickness is reduced, but that beyond a certain stage further rolling of the same material to very thin gauges may produce a clear surface. This is consistent with other indications that the original defects from which spills are formed are largely internal and exist in the outer regions of the ingot, and that these are progressively opened to the surface and may be gradually obliterated by continued rolling and annealing.

As regards composition, there is much works evidence that the higher the copper content of brass, the more difficult it becomes to avoid spill defects. It is considered that the higher melting point of the higher copper brasses is mainly responsible for this effect, though the increase in the range of temperature over which freezing occurs may be also a factor of importance.

Microscopic examination of sections cut from defective strip shows spill defects to consist of fissures lying close and roughly parallel to the surface, and open to it at one or more points. Examples are shown in micrographs Fig. 19. In most instances foreign matter, probably oxide, is included in the fissure. Many spills consist of multiple parallel fissures, while others are ramified and variable in width. The general character of the defects, of which a large number have been examined, confirms the view that they are mainly the product of pre-existing internal closed fissures near the surface of the strip which tend to break open during rolling when the overlying layer has become sufficiently thin. It is frequently possible to detect in strip hidden defects of the same kind which have remained unopened to the surface, and are consequently not able to be detected externally: these would have formed the source of open spills at a later stage had the strip been further rolled. Sections showing defects of this character are illustrated in the micrographs Fig. 20.

Blisters.—This form of defect, consisting of local swellings showing on one or both sides of the strip, appears in three varieties in rolled brass. Strip in the “as rolled” condition rarely shows surface blisters, but on deforming the metal by bending, stretching or spinning, small elongated blisters may appear on the surface, showing that local planes of discontinuity exist just below the surface layer. These represent the stage immediately prior to their development into visible spills by breakage of the overlying film, and such sub-surface defects have been already described. On opening blisters revealed by deformation, the interior is frequently dark in colour and in certain cases has been found to be covered by carbonaceous matter. Many opened blisters have been found, however, to be clean internally and others to be only faintly oxidised.

A second and different type of blister appears on both surfaces

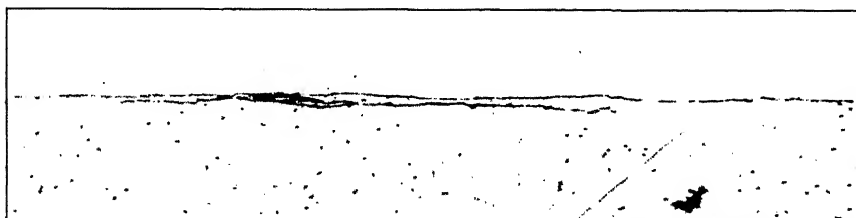
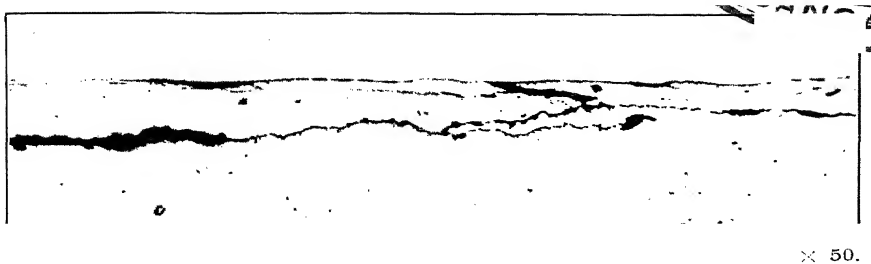


FIG. 19. $\times 100$.
Microsections of typical "spill" defects in rolled brass strip.

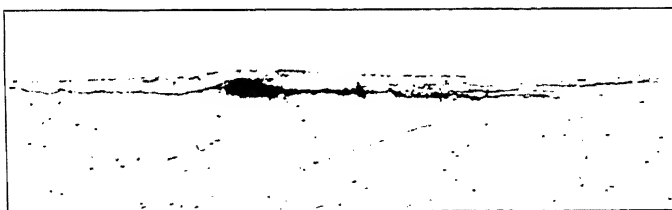
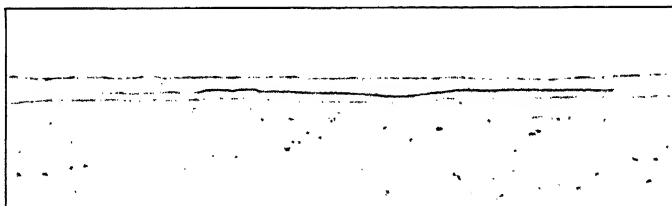


FIG. 20. $\times 100$.
Microsections showing internal defects in rolled brass strip not open to surface.

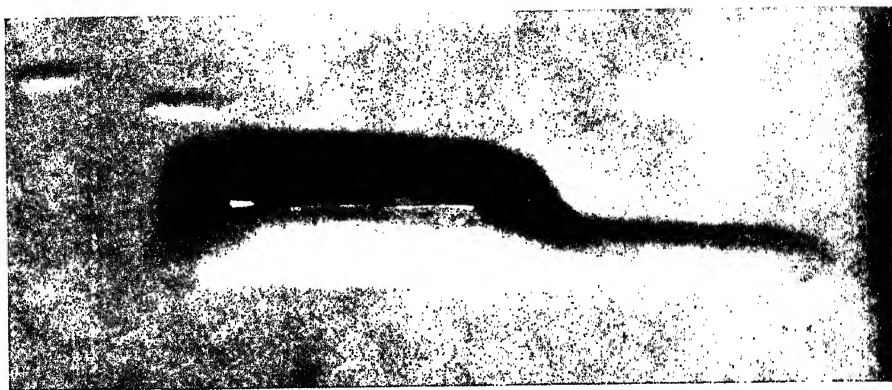
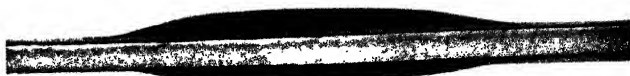


FIG. 21.

Abnormally large blister in rolled brass strip. Scale $\frac{1}{2}$ full size.

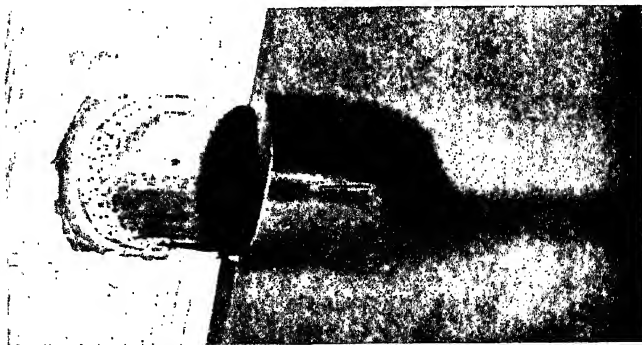


FIG. 22.

Interior of large blister illustrated in Fig. 21.

of the strip after annealing; on re-rolling it flattens and is not again disclosed without further annealing. An abnormal example of a defect of this type is shown in Fig. 21 and was specially selected for investigation in order to examine the gas enclosed. The blister had occurred during the hot rolling of brass containing 59.7 per cent. copper, 0.42 per cent. lead and 0.002 per cent. phosphorus. The bulge amounted to 0.268 inch on one face and 0.206 inch on the other, of strip 0.28 inch thick. By drilling both sides of the blister under water, the whole of the gas was collected (33.5 c.c. at N.T.P.). The internal volume of the blister was found to be 21.0 c.c.; hence the pressure of gas at 0° C. in the blister before opening was 1.6 atmospheres. Analysis of the gas showed it to consist of—

Carbon dioxide	CO ₂	0.24 per cent	
Carbon monoxide	CO	0.36	„ „
Oxygen	O ₂	2.29	„ „
Nitrogen	N ₂	6.62	„ „
Methane	CH ₄	2.07	„ „
Hydrogen		88.40	

The interior of the blister (see Fig. 22) was covered with a smooth yellowish film of oxide with dark brown bands near the edges and streaks of white zinc oxide. It is evident that the blister had originated in a central gas cavity in the ingot and had been formed, during hot rolling, when the containing walls became too thin to withstand the internal gas pressure. The composition of the gases suggests that they had been entrapped in the ingot during casting, as water vapour, hydrogen and hydrocarbons. The brass had been electrically melted and the only possible source of such gases would appear to be the oil used in the coating of the mould (water-cooled copper).

Blisters of similar characteristics and origin to the example described, but very much smaller (generally less than 1 inch in length), are most liable to occur in large ingots of low-copper brasses. Under certain conditions central blisters distended on both faces are also encountered on annealing strip cold-rolled from ingots of high-copper brasses, *e.g.* gilding metal; these may possibly be due to the entrapping in the central portion of the ingot of gases evolved from the molten metal on solidification.

A third type of blister is found in cross-rolled sheet, and is a common source of difficulty in manufacture. The defects appear as small roughly circular blisters distributed at random on the sheet, and the interior surfaces are almost invariably clean. From the amount of working undergone by the ingot before the appearance of blisters it is clear that the original defects from which they were formed must have been of extremely small dimensions. Representative examples of such sheet blisters are shown in Fig. 23, and in section in Fig. 24. The character of these defects is different from that of the previously described type of blister found on strip, since any one blister appears on one face only and ingots which give strip free from blisters of the first two types are often found to show small blisters of the third type when cross rolled.

Lines.—On pickling samples of strip, straight dark lines of varying length resembling shallow grooves may appear on the surface. These have been considered to be due to elongated inclusions of non-metallic matter lying under the surface and rendered visible by removal of the covering layer during pickling. An example is shown in Fig. 25. It is unlikely that this type of defect would be the cause of much difficulty during polishing except when present in a large amount. Microscopic examination of strip sectioned longitudinally at various stages of rolling indicates that non-metallic inclusions are first, to some extent, flattened and elongated but later break up into separate fragments, forming strings which lengthen as the rolling progresses. A typical series of strips from the same ingot showing this progressive action is illustrated in Fig. 26. The microscopic evidence available suggests that the metal separating the particles is not completely welded by cold rolling and that strings of included material would probably behave as a continuous plane of weakness in their effect on the mechanical qualities of the strip. This view is in agreement with the lower absorption of energy in Izod impact tests taken transversely to the direction of rolling, in relation to the results given by longitudinal tests.

The observations which have been made on various samples of strip would suggest, however, that on the whole, non-metallic inclusions are not an important source of weakness or defectiveness in brass strip. The degree to which inclusions are present



FIG. 23.—Small blisters in cross rolled brass sheet. $\times 3$.



FIG. 24.—Section through small blisters in brass sheet. $\times 50$.

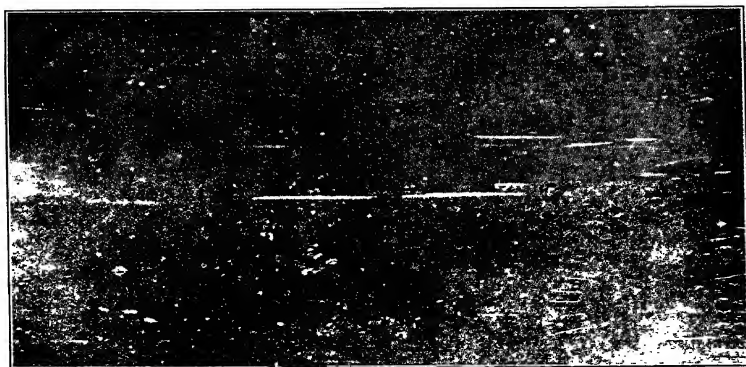
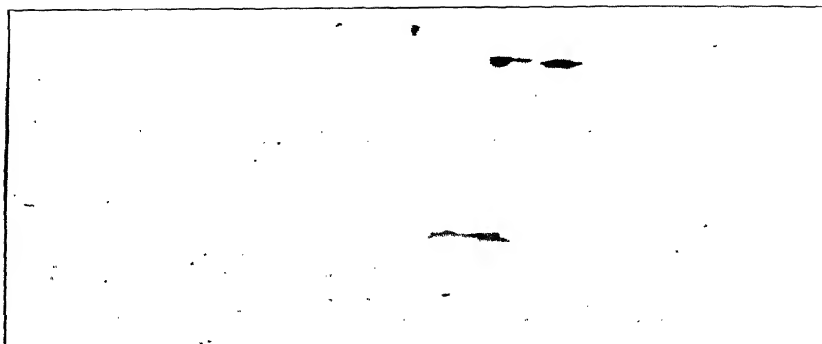


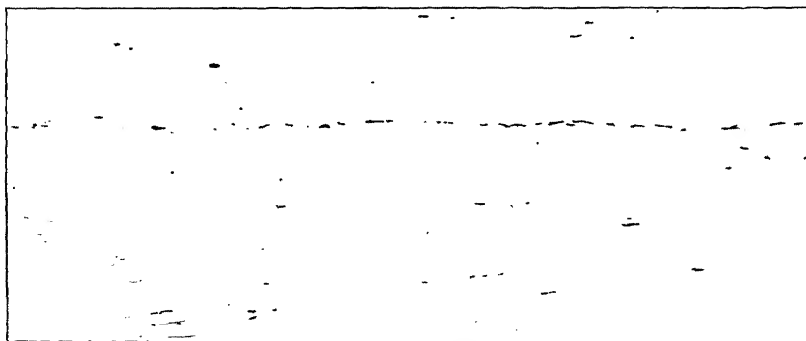
FIG. 25.—Fine longitudinal marks on rolled brass strip due to non-metallic inclusions. Actual size.

PLATE XVIII.



Reduction 50%.

Reduction 94%.



Reduction 97%.

FIG. 26.

Microsections illustrating behaviour of non-metallic inclusions on rolling. $\times 100$.

in brass is usually much smaller than in many other commercial metals and alloys.

Laminations.—In addition to subsurface discontinuities such as incipient blisters, a further type of defect which may be referred to as “laminations” may occur in strip, generally in the central region. These are comparatively rare in 70/30 brass, and are found mainly in brasses of lower copper content. They appear to be due to clean cavities which are not welded by cold rolling—probably contraction cavities. The shearing of discard, particularly from the end of the strip representing the top of the ingot, reveals defects of this type as lines on the shear cut, often near the middle of the thickness. Fig. 27 illustrates a fracture through strip showing such defects. It is infrequent that large rejections are made owing to small laminations, and in general their presence may be assumed to cause little difficulty in manufacture, except at the ends of the walls of drawn articles where, in metal of thin gauge, buckling may result.

The fact that deformation of rolled brass is capable of distorting the surface layer sufficiently to reveal underlying defects suggests a number of tests of simple form which might be applied to detect unsoundness. By stretching, bending or spinning samples of strip to only a slight extent, it is possible to make a relative estimate of the degree of subsurface unsoundness present. An example of a defect revealed by spinning is shown in Fig. 28. Extensive working of the material is unnecessary and does not reveal any defects additional to those visible at an early stage of permanent deformation. The fact that composite strips of brasses of varying composition, when spun into cups, show no variation in the position or proportional thickness of the different layers, shows that spinning cannot produce defects by bringing to the surface deeply seated unsoundness. The existence in strip of lamination defects, including subsurface defects, may be quickly revealed by bending a strip after light pickling of the surface in nitric acid and immersion in mercury or mercurous nitrate solution. The brass rapidly cracks and breaks under a small stress, and laminations appear clearly visible on the fracture. This effect is due to the fact that the penetration of mercury under the influence of stress tends to be interrupted by discontinuities

and consequent changes in the distribution of stress causes the fracture to form a series of steps. A specimen tested in this manner is shown in Fig. 29.

From the foregoing descriptions of the features of ingots and of rolled strip, a useful degree of correlation is possible. Conclusions reached in this way should, however, be confirmed by direct experimental work in which the ingot as cast and the strips at successive stages of rolling are examined systematically. A convenient method is to divide the ingot longitudinally into halves, one half being sectioned and tested and the other portion rolled, a suitable length being cut from the strip at each stage of rolling and examined to provide a complete history of the material. The following summary connecting the qualities of ingot and strip is based on the results of an examination of this kind on a number of ingots.

In the manufacture of strip (that is, material which is not cross rolled) the most important sources of defects such as spills and blisters are surface irregularities and subsurface cavities in the ingot. In commercial practice the surface of the cast ingot is generally smooth and of high quality and local accidental defects occurring can be largely eliminated by skilful "chipping" at an early stage of rolling. Ingot surface defects not thus removed will generally give rise to spills in the rolled strip. Subsurface cavities in the ingot become evident on the surface of the strip at a stage depending on their nearness to the ingot surface and the degree of reduction in rolling, since the layer of metal overlying the cavity must be reduced to a very small thickness before fracture or distortion renders the defect visible. Should such a cavity remain close to the surface but undisclosed by rolling it may appear on further working of the material.

The very small distributed cavities in ingots rarely affect the quality of strip, but where they are extended in two directions, as in the cross rolling of sheet, they are the main source of blister formation. Small spills, not elongated, may result from the breaking open of small blisters on sheet, and may occur at any thickness.

In brass of low copper content (below about 63 per cent.) and also in the copper-rich alloys such as "gilding metal," cavities of relatively large dimensions containing gases may be formed

PLATE XIX.



FIG. 27.

Fracture through rolled brass strip containing central defects.

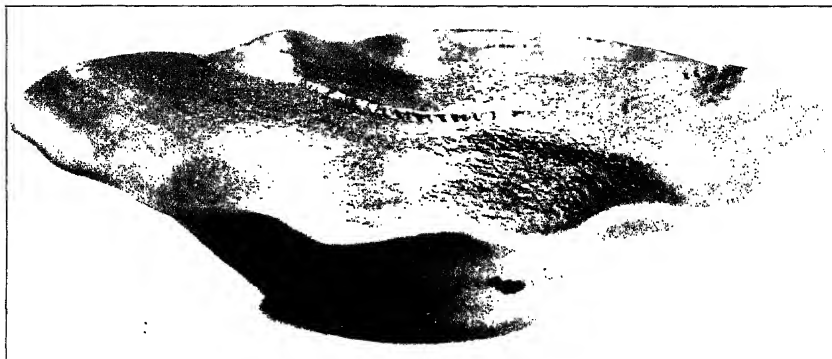


FIG. 28.

Previously invisible spill in 70-30 brass strip revealed by spinning.

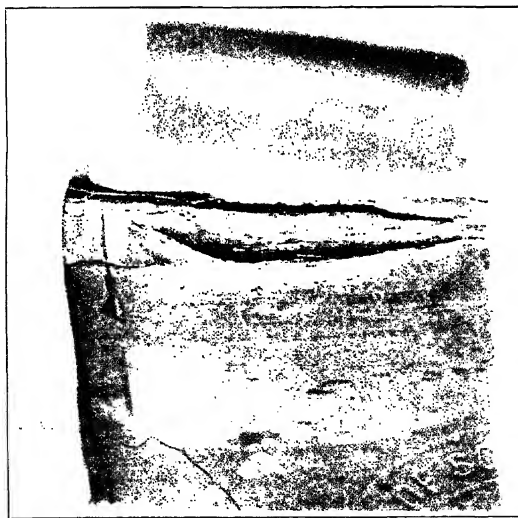


FIG. 29.

Lamination in rolled brass strip revealed by mercuric nitrate test.

near the centre plane of the ingot. These are potential sources of blisters which form a bulge on both faces of the strip. Distension of the blister occurs on annealing due to gas pressure. The defect is probably only a minor trouble in manufacture owing to the infrequent occurrence of large central cavities containing gas.

Contraction cavities, where distributed over the central region of the ingot, as in the type of unsoundness known as "sponginess," do not appear to influence the behaviour of the metal in rolling. Only when they occur in high concentration on the central plane of ingots, particularly those of low copper content (near 60 per cent.) is there evidence of central weakness of the rolled strip.

Inclusions of non-metallic matter in ingots, mainly zinc oxide, are similarly nearly negligible in their effect on the quality of strip, beyond causing a small variation in the transverse mechanical properties. They are formed during pouring, are generally small in amount, and do not segregate.

It is thus clear that cavities differing in type as regards size and distribution, but similar in the respect that they are all due to gas inclusions, constitute the most important defect of brass ingots from the point of view of the final quality of the material, and where, as is frequent in commercial practice, the standard of surface quality is high, subsurface unsoundness is considered to be the main cause of defective strip.

CHAPTER V

THE LIQUID METAL

Properties—Surface State—Effects of Added Elements.

BEFORE attempting an analysis of the casting process, it is necessary to take into account the conditions in the liquid metal when in the crucible ready for pouring, and their effect during the passage of the stream into the mould. The slag, which up to the time of removal of the crucible from the furnace forms a protective layer, is of minor importance, and is skimmed off or otherwise mechanically removed as the first step in the casting operations. It is clear that in brass casting, though not in the casting of all alloys, any slag poured with the metal tends to rise rapidly to the top, and it has previously been shown that non-metallic inclusions in cast brass are rarely large in amount. Slag, therefore, should not be a troublesome factor in brass casting, though lack of care in skimming may result in occasional local defects, mainly on the ingot surface.

It is known that the viscosity of all metals is low at all temperatures above the melting point, and from the practical point of view this property may be entirely neglected, since its variation would not affect the behaviour of the molten metal except under special conditions, such as where large quantities of included matter are held in suspension or where solid particles of metal are present due to the alloy having cooled below the liquidus temperature.

Similarly, the true surface tension of brass, although high (like that of all other molten metals), cannot be varied sufficiently to exert any influence on the casting behaviour of the metal comparable with that of other surface conditions which arise in practice.

Pure brass in the liquid state exposed to air is rapidly covered by a surface film of oxide, which is "wetted" by the brass and is closely adherent; it has the apparent effect of reducing the

PLATE XX.



FIG. 30.
Defects in surface of brass ingot due to film of oxide. $\times 2$.

evolution of zinc. When the film is broken, zinc vapour is more rapidly given off, but a new film is immediately formed over the disturbed area and evolution of zinc again largely ceases. It is thus evident that the influence of the oxide skin is mechanical. The film is not extensible compared with the liquid surface, and its presence cannot therefore be said to have the general effect either of lowering or increasing the true surface tension of the liquid alloy; with some methods of measurement the value may be lowered and with others increased. Its actual behaviour appears to be that of a more or less rigid casing, restricting movement near it and when broken being instantly repaired by fresh oxidation. In the case of metals it is convenient, however, to regard the effect as raising the apparent surface tension of the metal and impeding its smooth flow throughout the formation of the ingot. Casting defects commonly ascribed to variations in "fluidity" or viscosity are in many cases due to oxide or other films on the surface of the molten metal. It is evident from the appearance of surface defects in commercial brass ingots, and comparison of such defects with those in ingots intentionally poured under oxidising conditions (see Fig. 30), that the restriction of flow is the main cause of the bad effects of oxidation during the casting of brass. These include surface irregularities and folds due to successive waves of metal rising up the mould surface, partially welded globules of splashed metal, and internal inclusions of oxide film. A photomicrograph from near the surface of an oxidised ingot showing included films is reproduced in Fig. 31.

At the melting point of 70/30 brass the vapour pressure of zinc is low, but at temperatures of the order of 1100° C. such as are normally used in casting, the partial pressure becomes sufficiently high to cause disruption of the surface oxide film, and the rapid escape of zinc vapour. Oxidation of this zinc vapour occurs externally to the oxide film, producing free zinc oxide particles which are carried away in the thermal air currents above the melt or rest loosely on its surface, and do not adhere. Previous work ¹ has shown that this form of zinc oxide, not formed in contact with the metal, is not wetted by subsequent contact with molten brass and does not form a coherent film. When introduced into a melt and intimately

mixed it is not retained in suspension except under special conditions which are difficult to contrive even experimentally. It is clear, therefore, that assuming normally careful skimming, oxide formed externally in this way may be regarded as negligible from the point of view of ingot quality, and that in the avoidance of defects in brass caused by oxidation it is necessary to consider only the adherent surface film.

In practice, oxidation is largely avoided by the provision of a reducing atmosphere in and around the mould, and it is the custom in some foundries to place on top of the molten metal a block of wood which maintains a similar reducing atmosphere inside the crucible. Oxidation defects, therefore, are not of general occurrence except for local effects due to a momentary failure of the reducing atmosphere, such as may be caused by draughts in the foundry. These accidental defects may cause considerable trouble in manufacture, and for that reason the tendency is to employ a generous excess of volatile material in the mould for the provision of protective reducing gases. In other directions, however, this practice has definite disadvantages (which will be discussed later) and there is undoubted scope for other more reliable methods of dealing with oxidation difficulties.

The effect of added elements on the mode of surface oxidation of liquid brass may be inferred, with limitations, from a comparison of their heats of oxidation (calories per gram-atom oxygen) with that of zinc. Values for some of the common elements are given in Table II.

While the values given refer to ordinary temperatures, they may be regarded as a broad guide to the relative behaviour of the elements in molten brass.* It may therefore be expected that zinc would oxidise preferentially to all the elements shown below it, and that magnesium and aluminium would be the most effective in modifying the character of the surface of liquid brass in contact with an oxidising atmosphere; the effect of the non-metal phosphorus is specially referred to later.

* Little fundamental work has been done on the application of thermo-dynamics to the problems of deoxidation, etc., in non-ferrous metals, but it is interesting to note that in a recent paper on the deoxidation of liquid steel² the author concludes that the deoxidising powers of magnesium, aluminium, silicon, manganese and chromium in steel at 1600° C. are in the same order as that indicated in Table II by the heats of formation of the oxides at ordinary temperatures.

TABLE II
HEATS OF FORMATION OF METALLIC OXIDES

Element.	Oxide.	Heat of Formation of Oxide.	
		Calories per gram-molecule.	Calories per gram-atom of Oxygen.
Magnesium . . .	MgO	144	144
Aluminium . . .	Al ₂ O ₃	380.2	127
Silicon . . .	SiO ₂	193	95.5
Boron . . .	B ₂ O ₃	282.1	94
Manganese . . .	MnO	90.8	90.8
Chromium . . .	Cr ₂ O ₃	267.5	89.3
Zinc . . .	ZnO	85.4	85.4
Phosphorus . . .	P ₂ O ₅	369.4	73.9
Tin . . .	SnO ₂	137.8	68.5
Iron . . .	FeO	65.7	65.7
Nickel . . .	NiO	57.9	57.9
Copper . . .	Cu ₂ O	40.8	40.8

It would be anticipated that the presence in brass of an element more readily oxidised than zinc would result in the substitution of the zinc oxide film by a film of oxide of the added metal, and the main point of practical importance would be the relative strengths of the different films. If the new film were weaker than one of zinc oxide the effect of the additional element would be to reduce the risk of casting defects.

Aluminium is strongly disliked by foundrymen engaged in the crucible casting of brass and great care is taken to exclude any metal and scrap containing it from the materials used in manufacture. The presence of aluminium, even in small amounts, is easily detected on the top surface of an ingot by a distinctive metallic sheen and the absence of black or grey oxide. The effects of the addition of aluminium to molten 70/30 brass are considerable. Oxide of aluminium is formed, producing a thin surface film. Evolution of zinc is markedly reduced at a temperature of 1100° C. and almost ceases at 1050° C. Any zinc oxide produced by breakage of the film is formed externally to it and does not affect its character, or adhere to the surface. Ingots of brass containing aluminium, poured by the usual methods, show pronounced surface defects similar to those shown in Fig. 30, and the ordinary means used to prevent the oxidation of brass are ineffective when aluminium

is present. These facts indicate that the aluminium oxide film, though extremely thin, is much stronger than the zinc oxide film formed on pure brass. It is clear that the only satisfactory method of casting brass containing aluminium is one in which pouring is essentially a quiescent process not involving turbulence or pronounced motion in the mould, so that the film is not folded or rapidly broken.

The presence of magnesium up to 1 per cent. in 70/30 brass clearly results in preferential oxidation of the magnesium, since a stream of the alloy poured in air shows no evidence of zinc flame. The film formed is weak and ingots of magnesium brass, cast in air, are superior in surface quality to pure brass similarly cast. The material, however, is not free from oxide film effects and internal inclusions are numerous. These are evidently present in the melt, since the material is distinctly sluggish in casting. Addition of magnesium is, therefore, not a satisfactory means of dealing with the problem of the oxide film of molten brass.

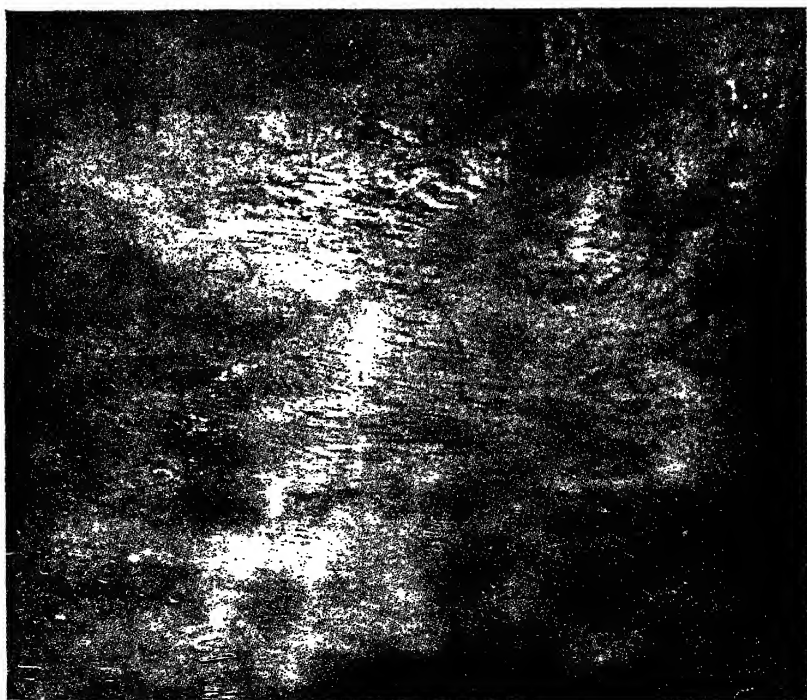
The effect of silicon is to produce a film, possibly of silicate, and silicon brasses are stated by Vaders³ to give extremely clean surfaces in chill castings.

The case of phosphorus is peculiar in that at the temperature of molten brass both the element and its oxide are gaseous. Some effect on the surface conditions is therefore to be anticipated, although the oxidation of phosphorus may not be preferential to that of zinc. Brasses to which has been added as small a quantity as 0.05 per cent. of phosphorus show a considerably greater zinc evolution than occurs with pure brass, aluminium brass or silicon brass, suggesting that any film formed is extremely weak. The zinc oxide produced is in the pulverulent non-adherent form. While the exact nature of the changes occurring at the surface of molten brass containing phosphorus in contact with air is not known, it would appear probable that a thin external layer of gaseous phosphorus or liquid phosphate is formed, through which zinc vapour can readily be evolved to burn on contact with the external atmosphere.

Casting experiments with brass containing phosphorus poured in air show that the full effect is obtained with as small a phosphorus content as 0.04 per cent. The molten stream is



FIG. 31.
Section near surface of oxidised brass ingot showing included oxide
films. $\times 50$.



continuously enveloped in zinc flame, but no adherent zinc oxide film is formed and the resultant ingot is smooth and free from surface defects (see Fig. 32). A peculiarity of molten brasses containing phosphorus is the pronounced tendency of the material to splashing, owing to the fact that the apparent surface tension is not increased by a restrictive film. The disadvantage of splashing occurring within the mould would be expected to be counterbalanced by the ease with which the oxide-free metal is welded into the main mass.

The observations such as have been described give only a broad impression as to the resultant effects of the added elements. Direct measurement of the strength of the surface films is a matter of considerable difficulty, but in brass, use may be made of the convenient fact that one of the mechanical effects of the oxide film is to retard the escape of zinc vapour from the metal. Consequently a relative quantitative estimation can be made of changes in the strength of the surface film on addition of other elements, by measuring the zinc evolution under controlled conditions. The experimental method utilised by the authors consists in heating about 30 grams of the alloy at the required temperature in a tube furnace through which air is passed in a regular stream for 30 minutes. Nitrogen containing a small percentage of oxygen may be used in order to provide an atmosphere permitting the evolution of zinc through a relatively thin oxide film, and hydrogen to show the evolution of zinc from the alloy in the absence of any protective film. After solidification the metal is lightly pickled to remove adhering oxide and again weighed. The results of some experiments of this kind on 70/30 brass are given in Table III.

As would be expected, the evolution of zinc is distinctly greater in liquid brass than in solid brass at 800° C. The thinner film formed in nitrogen containing a low concentration of oxygen permits further increase to an amount which is, however, much less than that obtained when formation of a film is totally prevented, as by heating in hydrogen. The effect of a small quantity (0.2 per cent.) of aluminium is to strengthen the film to some extent, the evolution of zinc being distinctly reduced in all circumstances where oxygen is present.

With further addition of aluminium to 2.5 per cent., the oxide film is strengthened sufficiently to reduce zinc loss to a

negligible amount, even when the concentration of oxygen present is very small. The addition of silicon gives results similar to those for pure brass. Molten brass containing phosphorus loses zinc at a rate which is practically independent of the nature of the atmosphere in which the brass is heated. The alloy may thus be assumed to oxidise without the formation of any continuous surface oxide film, and in the casting of ingots, such conditions render it unnecessary to provide any means for the avoidance of oxidation, since the ill effects due to restriction of flow cannot occur when phosphorus is present.

TABLE III

OXIDATION AND ZINC LOSS OF 70/30 BRASS CONTAINING ADDED ELEMENTS, HEATED IN AIR, NITROGEN AND HYDROGEN FOR 30 MINUTES

Alloy.	Treatment. ° C	Loss, %.
70/30 Brass	800° in Air	0.25
	1050° in Air	1.7
	1050° in Nitrogen (+ 2% Oxygen)	4.7
	1050° in Hydrogen	12.6
Aluminium Brass (0.2% Aluminium)	800° in Air	0.27
	1050° in Air	0.92
	1050° in Nitrogen (+ 2% Oxygen)	2.52
	1050° in Hydrogen with flux	13.8
Aluminium Brass (2.5% Aluminium)	800° in Air	0.003
	1050° in Air	0.007
	1050° in Nitrogen (+ 2% Oxygen)	0.071
	1050° in Hydrogen	15.3
Silicon Brass (2.0% Silicon)	800° in Air	0.12
	1050° in Air	0.67
	1050° in Nitrogen (+ 2% Oxygen)	6.0
	1050° in Hydrogen	12.8
Phosphorus Brass (0.05% Phosphorus)	800° in Air	0.36
	1050° in Air	15.1
	1050° in Hydrogen	16.4

These observations apply to brasses which, apart from the addition of the single elements stated, are of high purity. Phosphorus does not necessarily prevent the formation of an oxide film in brasses containing other elements, and if phosphorus and aluminium, for instance, are present together in a

brass, any effect of the phosphorus on the metal surface is entirely masked by the aluminium. The affinity of aluminium for oxygen is at these temperatures so much higher than that of phosphorus that a phosphorus-aluminium brass behaves in the same way as a pure aluminium brass.

The effects of additions of aluminium and of phosphorus on the mechanical and physical properties of copper-zinc alloys in the solid state are dealt with in Appendices C and D, respectively (pp. 164 and 176).

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CHAPTER VI

STREAM TO INGOT. INGOT STRUCTURE

Conditions in the Mould during Casting—Experiments on Ingot Formation.

UNDER ideal conditions an ingot just cast may consist of a full mould of liquid metal at a uniform temperature, and the known physical laws may be used to explain the results of any applied process of solidification.* The ordinary casting methods, when used in conjunction with sand or other refractory moulds which extract heat slowly from the ingot, provide some approach to uniformity in temperature, but in chill castings made in moulds of conductive material the rate of cooling involved must clearly produce extreme variations. It is mainly this effect which causes different methods of casting to produce ingots of different character in various respects, such as structure and distribution of unsoundness. Apart from the effects of the properties of the molten metal, therefore, the history of an ingot must be regarded as commencing with the first drop of metal entering the mould. From these considerations it will be clear that the flow of the metal within the mould is an important feature of the course of solidification, and that the study of ingot building is likely to be a valuable source of knowledge on which to base comparisons of methods, as well as deductions regarding the formation of casting defects.

The path of a stream of liquid being poured into a mould can be illustrated by pouring successive amounts of coloured liquids into a mould of glass. Such illustrative methods are,

* These conditions are practically attainable with low-melting-point metals, including aluminium, with the aid of specially devised apparatus, and are in fact the essential basis of the recently developed R.W.R. casting process.¹ With alloys of relatively high melting point the technical difficulties to be overcome are very great, and at present such a process is not industrially feasible for brass and many other common alloys.

however, unsatisfactory in that they take no account of the important factor of cooling proceeding concurrently with pouring, even when waxes or low-melting-point metals are used. Experiments on the actual type of material being studied are therefore in this case, as in many others, the only satisfactory means of obtaining reliable knowledge.

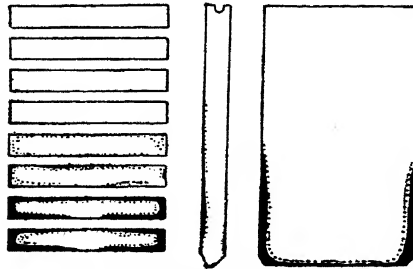
The method used by the authors ² was the preparation of ingots from two alloys, similar as regards freezing point and density, but possessing different colours. These requirements are most nearly met by a copper-zinc alloy containing 90 per cent. copper, almost copper-red in colour, and a "nickel silver" containing 50 per cent. copper, 30 per cent. zinc, and 20 per cent. nickel, a practically white alloy. The liquidus is 1050° C. in each case and the solidus approximately 990° C. and 950° C. respectively. Two crucibles are employed, containing the two alloys at a temperature of 1200° C. Suspended above the mould is a carborundum trough or tundish, 6 inches long, in one end of which is fixed a circular nozzle through which the metal passes into the mould. The two crucibles are held, one on each side of the trough, and the brass is poured first, the "nickel silver" following before the last of the brass leaves the trough. In this way ingots can be prepared from an unbroken uniform stream of metal, while the alloy comprising the stream can be changed at any prearranged stage by varying the amounts of metal in the two crucibles. Different sets of casting conditions covering most of the methods used industrially can be examined in this way, and for each set of conditions it is advisable to make three ingots containing the two alloys in the proportions 25:75, 50:50, and 75:25 respectively. The ingots are sectioned by cutting a longitudinal central slice, and making transverse cuts dividing the remainder into eight equal parts. On polishing and etching the sections, three zones of colour, red, yellow and white, are distinguishable, the yellow zone (and possibly a portion of the adjacent white zone) representing regions in which mixing of the liquid alloys has taken place. It may be mentioned that reproduction of sections of composite ingots photographically is not generally satisfactory, in view of the difficulty of etching suitably to show clearly the different gradations of colour, and scale drawings have been found the most useful method of

illustration, the red alloy being shown black, the mixed region dotted, and the white alloy regions left blank. By this method of reproduction it is possible to add a section through the central plane at right angles to the central slice, constructed from measurements on the central slice and the transverse sections. Figs. 33 to 37 show the distribution of the two alloys in a series of ingots illustrating the effects of variation in the casting conditions, which are described and discussed below. All the ingots shown were made in cast-iron moulds of dimensions $12 \times 6 \times 1$ inch, the rate of pouring being equivalent to a rise of metal in the mould of about 1 inch per second, similar to that used in ordinary practice.

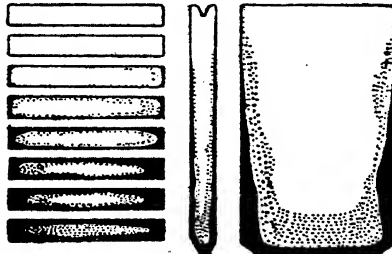
Fig. 33 represents ingots made in a vertical mould, the stream of metal being directed down the centre, avoiding the mould faces. Solidification was complete within a few seconds after completion of pouring. It is apparent from these results that the motion of the metal within the mould produces a formation of the ingot and resulting mode of solidification markedly different from the conception which has previously been widely held. No evidence is seen of the filling of the mould by successive layers. Particularly striking features are that the metal poured in after the mould is three-quarters full penetrates nearly to the bottom of the ingot, and that traces of the metal forming the first 50 per cent. of the melt poured are found adjacent to the faces of the mould at the top of the ingot.

Reconstructing the sequence of events in the formation of the ingot, it would appear that the process is one of continuous displacement, occurring concurrently with the rapid extraction of heat from the surfaces. The metal first entering the mould commences to solidify and the temperature of the liquid central region falls towards the freezing point. This region is penetrated by the incoming stream, and the metal formerly occupying it is displaced upwards and outwards to the faces of the mould. Some mixing of the metal in the stream with the cooler metal in its path occurs, and this mixture, cooler than the stream, is in turn displaced in the same manner as pouring proceeds, forming a rapidly cooled outer layer or skin, but solidification inwards from this layer must be continually hindered by the constant penetration of hot liquid metal.

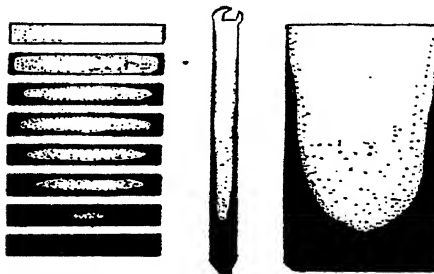
By casting under conditions similar to those just described, but with the mould leaning at an angle of 15° to the vertical



TOP CAST. MOULD VERTICAL. BRASS 25%
SINGLE STREAM DOWN CENTRE.



TOP CAST. MOULD VERTICAL. BRASS 50%
SINGLE STREAM DOWN CENTRE



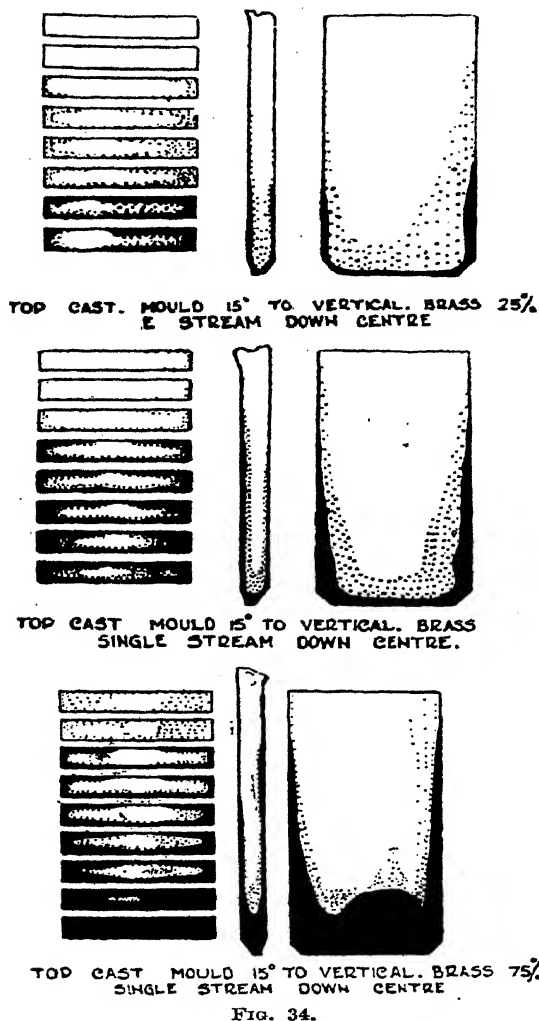
TOP CAST. MOULD VERTICAL. BRASS 75%
SINGLE STREAM DOWN CENTRE.

FIG. 33.

as is frequent in industrial practice, the distribution shown in Fig. 34 is obtained. While in the transverse sections there is no great difference from the former series, the longitudinal

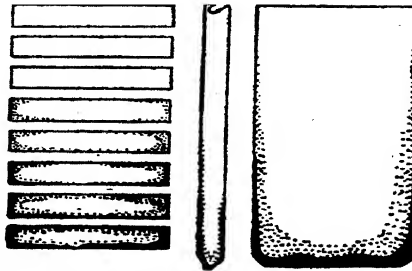
THE CASTING OF BRASS INGOTS

section shows that the path of the stream down the back surface of the mould is occupied by a portion of the last metal poured,

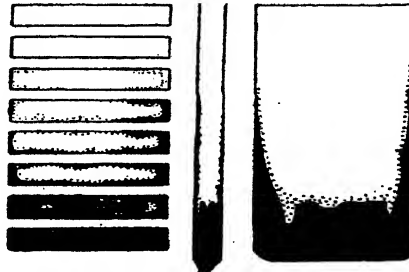


and the displacement of liquid up the surface of the mould by the stream is more marked on the opposite face. There is clear evidence, therefore, that in this method of casting the

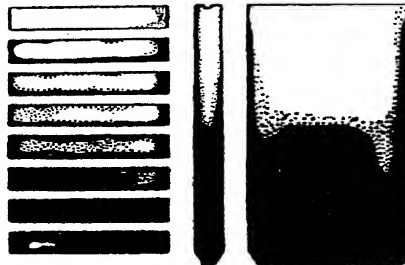
back face of the mould is continuously swept by the stream, and that until the completion of pouring little or no solidifica-



TOP CAST. MOULD VERTICAL. BRASS 25%
TUNDISH WITH 4 HOLES



TOP CAST. MOULD VERTICAL. BRASS 50%
TUNDISH WITH 4 HOLES.



TOP CAST. MOULD VERTICAL. BRASS 75%
TUNDISH WITH 4 HOLES.

Fig. 35.

tion occurs in that region. At the opposite face, however, cooler metal is being supplied from the lower portion of the ingot and solidification may occur earlier. In this way the

ingot is likely to solidify unsymmetrically across the thickness, while the front and back portions of the mould will be heated

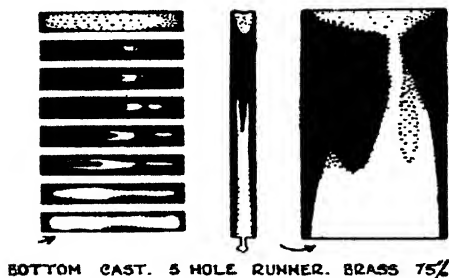
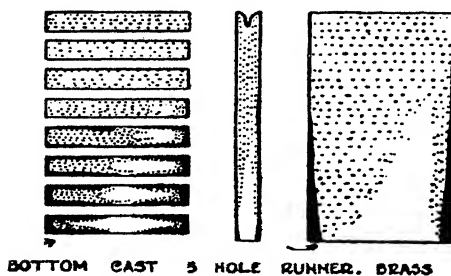
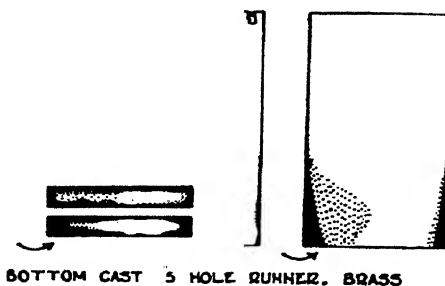
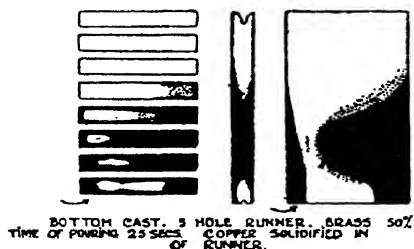


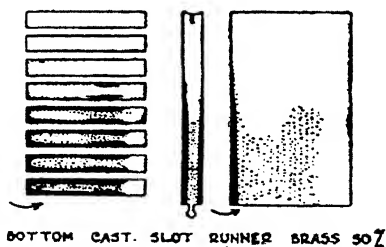
FIG. 36.

to considerably different temperatures. It will also follow that any defects due to the course of the stream will be unsymmetrically distributed.

Fig. 35 shows ingots cast, again under the same conditions, in a vertical mould, but with a pouring trough provided with four small holes spaced along its length in place of the single nozzle. This method of pouring has been widely used in conjunction with large electric melting plants and is now being adopted, in view of the results of research on the subject, by some crucible casting foundries. Since the division of the stream into a number of smaller streams must reduce the impetus of the liquid metal as a whole, the method would be



(a)



(b)

FIG. 37.

predicted to be beneficial in reducing the temperature variations due to displacement effects. This influence is confirmed by the composite ingots illustrated in Fig. 35, which show distinctly less penetration of the liquid in the mould by the incoming stream. Although the type of ingot formation is generally similar to that produced by a single stream, it appears justifiable to assume that the trend of solidification in the lower portions of the ingot will not be so much influenced by the addition of the hot metal in the upper portions.

Fig. 36 shows a series of ingots which were bottom-cast,

using an extemporised vertical fireclay runner or "trumpet" and a horizontal runner of the same material provided with five holes. The formation of the ingots is similar to what might be predicted from the results of the former top-poured ingots, assuming the impetus of the stream to be somewhat reduced by friction. The first portion of the metal to enter the mould remains to a certain thickness on the mould faces, and the remainder is mixed with the incoming stream. Penetration by the stream and further mixing continues as pouring progresses, and the last 25 per cent. of metal poured forms the middle region and part of the top of the ingot. Fig. 37(a) shows an ingot cast too slowly. In this case the first poured red-brass alloy solidified in the far end of the runner, choking two of the holes, and the resulting local concentration of the stream resulted in complete penetration of the first poured (lower) half of the ingot by the latter 50 per cent., which ultimately formed nearly the whole of the top half of the ingot. In an ingot cast with a bottom runner having a slot in place of holes the stream is again concentrated, but at the end farthest from the entrance to the runner (Fig. 37(b)). Mixing takes place to a considerable extent, but the top of the ingot is in this case also formed from the last 50 per cent. of metal poured.

In a bottom-cast ingot the liquid surface rises smoothly without the projection of metal up the faces of the mould. There is no direct vertical displacement of the earlier poured metal, and it would appear that under certain conditions, bottom casting can give a much nearer approach to equalisation of temperature in the ingot than the usual methods of top casting. In this connection, however, the importance of the disposition of runners in bottom casting is emphasised by the results shown.

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CHAPTER VII

SOLIDIFICATION OF THE INGOT

Contraction—Feeding—Influence of Pouring Conditions on Contraction Cavities—Formation of Crystals—Influence of Pouring Conditions on Ingot Structure—Effect of Macrostructure on Rolling.

THE manner in which a mass of liquid metal solidifies when heat is conducted from its surface is well known in broad principle. This, in conjunction with knowledge regarding the formation of the ingot from the stream, provides a fairly complete conception of the casting operation and the probable character of the solid ingot which results from any particular type of process.

When a mould is filled with liquid metal heat is dissipated from the metal by conduction to and through the mould, and temperature gradients are set up. A solid envelope of metal is formed in contact with the mould and progressively thickens until solidification of the ingot is complete. Assuming uniformity of casting conditions and of temperature in the liquid, the surface of solidification moves inwards parallel to the walls and final solidification occurs midway between opposite cooling surfaces. During cooling, contraction takes place both in the liquid and solid metal, while in most alloys there is, in addition, a further larger contraction accompanying the change from the liquid to solid state.* Inside the initially formed outer

* The total volume change occurring in a metal or alloy in cooling from the liquid state to atmospheric temperature is the resultant of three separate effects: (a) the change in volume in the liquid before reaching solidification temperature, (b) the volume change on solidification, and (c) the volume change in the solid from the freezing point to room temperatures. Values for these changes have been obtained with accuracy for low-melting-point metals and alloys,¹ but there is considerable difficulty in obtaining reliable values for alloys requiring the use of high temperatures. From the published work of Bornemann and Sauerwald,² Endo,³ Saeger and Ash⁴ and Pilling and Kihlgren,⁵ it would appear that for copper and copper rich alloys the contraction in volume of the liquid on cooling through 100° C. down to the melting point is of the order of 1.5 per cent. and the contraction in volume on

layer of solid, therefore, an appreciable total contraction of the ingot occurs before it finally solidifies. Consequently, as the solid layer thickens during solidification of the ingot, the surface of the central liquid falls progressively, replacing the contraction in the lower portions and leaving a depression or "pipe" at the top. In a chill-cast thin ingot, however, solidification takes place to some extent during the filling of the mould. The total contraction is thus less than under the former conditions, and the surfaces of solidification are not truly parallel to the outline of the ingot. In spite of this, the central portion of the pipe, if allowed to form, is likely to extend deeply into the ingot as a result of the final contraction of the last central column of liquid. In crucible-casting practice the top surface of the ingot, during solidification, is maintained level by small additions of liquid, and the pipe is thus avoided. Should "feeding" be too intermittent, the added metal may fail to penetrate to the bottom of the pipe, leaving a hidden cavity, and the most reliable method of feeding consists in the use of a heated feeding head or dozzle of fire-clay, of suitable capacity, placed on the top of the mould before casting. This is filled to the top in casting and forms a reservoir of liquid which automatically and efficiently feeds the contraction of the ingot. The method is now being increasingly adopted in non-ferrous foundries and is most valuable where large quantities of metal are cast by mechanical means.

Perfectly complete feeding of the ordinary type of strip ingot is, however, almost impracticable. The small central column of metal which solidifies last can rarely be completely sound, since the downward flow due to contraction in the final stages is impeded by "bridges" of solid which tend to form across it, and solidification below such bridges leaves small cavities known as contraction cavities. It follows that contraction cavities are most liable to form in long thin ingots and least in short wide ingots, though the character of the alloy

solidification 4 to 5 per cent. Subsequent contraction in the solid on cooling down to room temperature may be 5 to 6 per cent. corresponding to a linear contraction of about 2 per cent. In ordinary foundry practice it is customary in pattern-making to allow 1 to 2 per cent. in linear dimensions for shrinkage of the casting in the mould, though this figure naturally varies not only with the alloy used but with the size and shape of the casting.

has a considerable influence. This is seen in the ingots described in Chapter III, which illustrate the effect of variation in copper content on the unsoundness of cast brass.

Contraction cavities are always present in metal which solidifies out of contact with a further supply of liquid, and can only be avoided by progressive solidification in one direction such as from the bottom upwards. These conditions need only apply, in an ingot, to the central column of liquid, and since in a strip ingot solidification progresses to some extent during the casting operation, some approach to upward solidification might be expected. Actually, however, the mode of formation of the ingot is such that this effect cannot be generally obtained in practice.

The building of the ingot by the liquid stream, as has been described, involves a displacement action which necessarily affects the temperature gradients within the liquid ingot. In a top-poured ingot cast by a single stream the metal first poured is rapidly cooled by the mould. A portion is solidified, while the remainder is gradually displaced up the walls of the mould by the incoming metal. During this displacement further cooling proceeds, and solidification of the ingot surface layer must therefore be extremely rapid. The constant penetration of the central region by hot liquid, however, causes the progress of solidification to be held up until the mould is completely filled. The ingot then consists of an outer shell with an interior of liquid metal roughly uniform in temperature, which therefore cannot solidify free from contraction cavities. Movements of the stream from front to back or side to side of the mould may be assumed to interfere locally with the growth of the solid shell, and cause corresponding irregularities in its thickness, which increase the liability to bridging. Should bridging occur near the top of the ingot a large contraction cavity (secondary pipe) would occur, but this is not frequently found, since bridging near the top is prevented by the hot metal added in feeding.*

The conditions in a top-poured slab ingot thus represent little approach to solidification from the bottom upwards,

* If allowed to form, such cavities may contain reducing gases, such as hydrocarbons from the mould atmosphere, and are considered to be the probable cause of large blister defects such as the example described on p. 29.

and may be regarded as favourable to the formation of contraction cavities, though not of primary or secondary piping. The action of feeding is, on the other hand, beneficial, tending to minimise contraction cavities and ensuring soundness of the top portion of the ingot. Quantitative determinations of the influence of variations in casting conditions and of feeding are dealt with fully in a later chapter.

The use of an intermediate pouring bowl or tundish having a number of holes reduces the impetus of the metal entering the mould by decreasing both the height of fall and the section of the stream. In these circumstances the degree of penetration occurring in the liquid ingot is correspondingly smaller, the solidification of the lower portion of the ingot is less delayed, and the ingot is less unsound.

The influence of the stream in pouring has another important aspect. It is well known that a stream breaking the surface of a liquid produces an injection effect (dependent upon certain properties of the liquid) which results in bubbles of gas from the surrounding atmosphere being projected below the liquid surface. This may be simply demonstrated in transferring any common liquid from one vessel to another. Although the gas appears to rise instantly to the top surface, this is so only with the larger bubbles; the smaller particles of gas rise at a lesser rate depending on their size. In the pouring of an ingot of brass the metal rising in the mould must therefore be visualised as containing large numbers of bubbles of gases from the mould atmosphere. Owing to the penetration of the stream, some of these will be carried down into the body of the ingot at the completion of pouring. While the largest rise quickly, the very small bubbles can escape only relatively slowly and are thus liable to be entrapped by the rapidly solidifying metal. This provides a satisfactory explanation of the general occurrence of numerous minute cavities in top-poured brass ingots, such as have been described previously (see p. 20). These defects, which are considered to be the direct cause of the formation of small surface blisters when brass is cross-rolled to sheet, are thus traceable as having been introduced by the action of the stream during the actual formation of the liquid ingot. Since any measure tending to reduce penetration would be likely to be beneficial it may be inferred that the use of a tundish is,

from this point of view also, an improvement in detail, though the effect may not be very great.

Bottom casting would appear likely to reduce piping, since the whole of the liquid within the mould on completion of pouring is probably at a uniform relatively low temperature, and the amount of liquid contraction taking place is correspondingly small. Feeding can also operate to some extent under suitable conditions from both bottom and top of the ingot. The smooth rise of the metal in contact with the mould is an important feature of bottom casting, and conduces towards perfection of surface quality of the ingot. It would also be predicted that bottom cast ingots, made with special provision for avoiding injection effects at the runner, would not contain minute gas cavities, and would be free from blisters on cross rolling to sheet. The bottom casting process is widely used in the steel industry, partly in view of the metallurgical advantages and partly as an economical means of casting under control a large volume of metal. Its application to brass manufacture and the difficulties involved are discussed later.

The crystal structure of ingots, being formed during the transition from the liquid to the solid state, provides a record of the combined effects of the conditions present during solidification. The crystal structure of metals and alloys in general and the changes which are brought about by variations in temperature and composition are dealt with in Appendix B on the Constitution and Density of the Industrial Brasses, p. 154.

Generally, the size of the primary crystals or dendrites in ingots renders it possible to view large areas of structure, and sections etched to reveal the macrostructure and any accompanying cavities form a valuable source of information. This manner of studying effects produced by variations in casting methods has already been extensively used in the large amount of experimental work by which the present high standard of steel casting has been attained, but in the wider field of non-ferrous metallurgy the large number of variables to be taken into account has placed many obstacles in the way of reconciling the results of the numerous separate investigations published. Although, broadly, the same factors must operate for all substances which crystallise on freezing, the differences among

metals and alloys in important characteristics (such as the melting point and range of freezing) cause considerable difficulties in the interpretation of ingot structure. In sand-casting, the high degree of heat insulation afforded by the mould considerably simplifies the conditions under which the metal solidifies, but the macrostructure of thin slab ingots cast in chill moulds is complicated by the rapidity of solidification.

Crystallisation of metal from the liquid state is only possible when a certain degree of under-cooling (*i.e.* a fall of temperature below the freezing point) has been attained. Through a certain range of under-cooling the change from liquid to solid can take place, but only when initiated by the presence of some of the solid phase or by mechanical shock; this is termed the "metastable" range. Below the lower limit of this range solidification may occur spontaneously without the initial presence of the solid phase; this is termed the "labile" range. In the macrostructure of metals and alloys two types of crystal are found. One type is often elongated in shape, formed by the successive deposition of solid on crystals composing the surface of solidification in the cooling mass; the other type is of approximately equiaxial (polyhedral) form built up on independent central nuclei. The thickening of a solid wall or shell enclosing a cooling liquid therefore requires an extremely small degree of under-cooling within the metastable range, while for the formation of equiaxial independent crystals one of two sets of conditions is necessary: either further under-cooling to within the "labile" range must occur initially, or crystallisation must be initiated by inoculation with solid particles or by shock, while under-cooled within the metastable range. Under the conditions present in the pouring of a metal casting, the origin of equiaxial crystals may be governed largely by the temperature range of solidification of the alloy; where a range exists (as, for example, in 70/30 brass) both modes of formation of independent equiaxial crystals appear to be possible, but where solidification occurs at one temperature (as in a pure metal and in 60/40 brass), formation can be due only to inoculation by the presence of solid particles in the liquid and in special circumstances to the effects of mechanical shock.

A section of a brass ingot illustrating the different types of crystal formation is shown in Fig. 38. When the mould is

PLATE XXII.

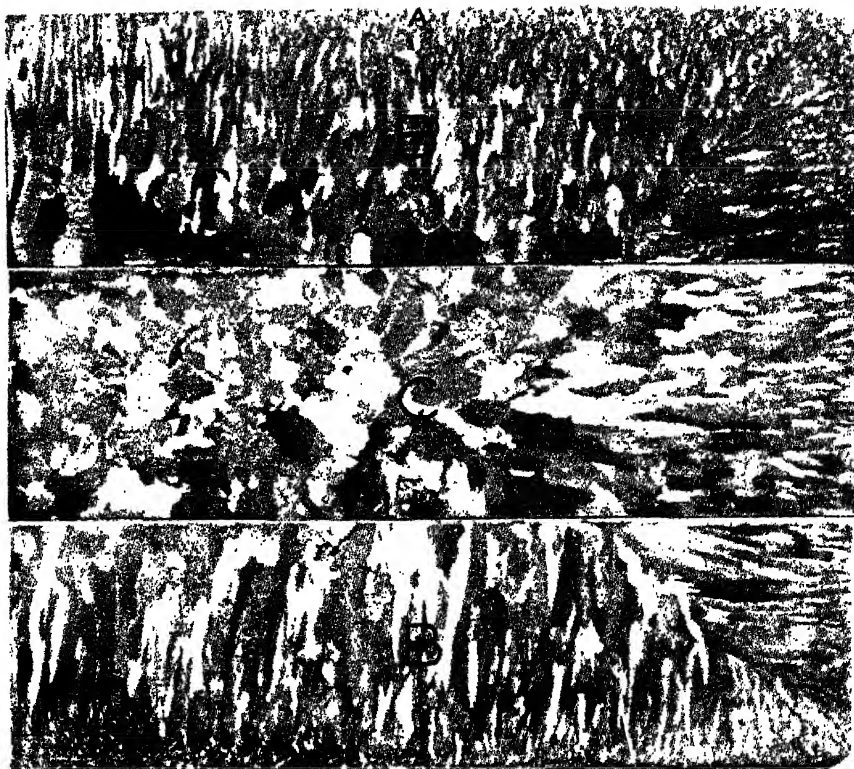


FIG. 38.

Section of brass ingot illustrating types of crystal formation. $\frac{2}{3}$ actual size.

filled with liquid metal, small crystals (A) are first formed in contact with the mould, and further crystals (B) grow inwards approximately at right angles to the surface. As long as the interior of the ingot remains completely liquid these continue to grow towards the centre, and, because each is confined laterally by other similar crystals, their shape becomes columnar. They may extend to the centre of the ingot or their growth may be limited by the formation of independent crystals (C) in the liquid interior.

The relative proportions of columnar and "free" crystals in a large steel ingot, which is slow in cooling, give fairly clear indications regarding the initial temperature of the metal and the rate of cooling, which latter is determined by the temperature, dimensions, and properties of the mould. In ingots cast in iron moulds the temperature gradient from centre to outside of the ingot tends normally to be steep. Consequently, columnar crystals are formed to some extent unless the casting temperature is very near the liquidus (upper limit of the freezing range) of the alloy. The higher the casting temperature the steeper is the temperature gradient, and the greater the depth to which the columnar crystals extend. Defects in steel ingots which are intimately connected with the macrostructure consist of irregularly shaped intercrystalline cavities due to shrinkage, and planes of weakness due to the solidification of low-melting-point impurities between the crystals; these also occur along diagonal junctions which are formed by the convergence of columnar crystals growing from adjacent faces of the mould, and at the bottom of the ingot on pyramidal planes. It is generally recognised that the most desirable macrostructure in steel is one consisting entirely of equiaxial (free) crystals, and accordingly the lowest possible casting temperature is frequently used, especially for some alloy steels, ingots of which are particularly liable to crack along diagonal planes during cooling.

Accepting that the same mode of solidification is followed by all metals, it might be assumed that the indications to be gained from large steel ingots apply directly to commercial ingots of non-ferrous alloys. This is true, however, only to a limited extent because of the smaller size and consequent rapidity of solidification of chill-cast non-ferrous alloy ingots and the

resulting strong influence which the flow of metal within the mould exerts on the mode of crystallisation. Particularly in thin ingots, local differences in conditions during cooling, which would be smoothed out in a large mass, are fixed as permanent structural variations.

The industrial brasses and many other non-ferrous alloys are normally cast at a temperature well above their melting points. A quietly and rapidly solidified thin mass of 70/30 brass, uniformly at an initial temperature of about 1100°C. , would consist almost wholly of columnar crystals extending from each face to the centre of the ingot. A similar result is obtained by casting highly superheated metal slowly with the minimum of disturbance of the liquid in the mould. In the ordinary method of casting, however, the penetration of the liquid ingot by the stream has the effect of stopping the growth of columnar crystals, which are separated at the completion of pouring by a zone of liquid cooled nearly to the freezing point during casting. At the same time, solid particles may be present in the liquid zone as a result of the injection of solid from the ingot top and erosion of the inner surface of the solidified layer. The tendency is therefore for crystallisation to develop throughout the central zone from independent centres, giving equiaxial crystals, and with them a certain amount of intercrystalline unsoundness, since the contraction cavities which form at each crystal boundary have no access to liquid except when near the top of the ingot.

Small contraction cavities are thus an inevitable sequel to the formation of equiaxial crystals, and form a zone of unsoundness ("sponginess") wherever this mode of solidification has occurred. Columnar crystals, which are necessarily in contact with liquid during the whole period of their growth, are consequently free from contraction cavities, except that, where the ingot consists entirely of columnar crystals, the central plane may be unsound due to bridging.

Ingots cast with a tundish are generally more columnar than when cast by a single stream, due to the smaller degree of penetration, and this structure corresponds to an increase in soundness.

The fact that turbulence is one of the principal governing influences in the formation of the ingot would strongly suggest

that in brass slab ingots cast with a mould coating of volatile material the added influence of this factor may be considerable. This subject is dealt with later in conjunction with the study of other properties of mould coatings (see Chapter X).

The extent of the range of temperature over which solidification of an alloy takes place is a further factor which appears to influence structure to a marked degree. An ingot of 70/30 brass, which has a relatively long range of solidification, may show a proportion of equiaxial crystals, while a similarly cast ingot of 60/40 brass, which has a very short range, consists largely of crystals of columnar form. This may be explained by reference to the fact that in an alloy having a temperature range of solidification, the crystals are not uniform in composition, but vary with the temperature during solidification. In the liquid portion comprising the central region of a solidifying ingot poured at a considerable superheat (or the whole of an ingot poured at a temperature very little above the freezing point), the temperature gradient may be assumed to be small. Since diffusion is slow relative to the rate of cooling, it is to be expected that adjacent to the growing crystals there may exist a layer of liquid of different composition and lower freezing point, while in the immediate region beyond it there is a sufficient degree of under-cooling to permit independent crystallisation. The greater the temperature range between the liquidus and solidus (beginning and end of solidification), the greater is the possibility of those conditions occurring, and the proportion of equiaxial crystals in an ingot is thus related to the range of freezing of the alloy. In the solidification of an alloy having such a freezing range (between liquidus and solidus), solid is being deposited at all temperatures within that range and it may be conceived that the surface of solidification lies between each of the two isotherms representing the liquidus and solidus temperatures respectively. To fulfil this condition the surface of solidification would most probably be of serrated form, as shown diagrammatically in Fig. 39 below, and the serrations would be dendrites projecting into the liquid to a depth depending on the steepness of the temperature gradient. Adjacent to the surfaces of the dendrites there would exist the layer of liquid of lower freezing point. There is in these circumstances no obstacle to the formation of independent

nuclei between the projections (at positions shown at A in the sketch), giving rise to a barrier of equiaxial crystals. In general, therefore, where an alloy solidifies over a range of temperature the conditions as a whole are conducive towards a structure composed partly of columnar and partly of "free" equiaxial crystals.

When the metal has no freezing range the surface of solidification would be conceived to be smooth, since the isotherms

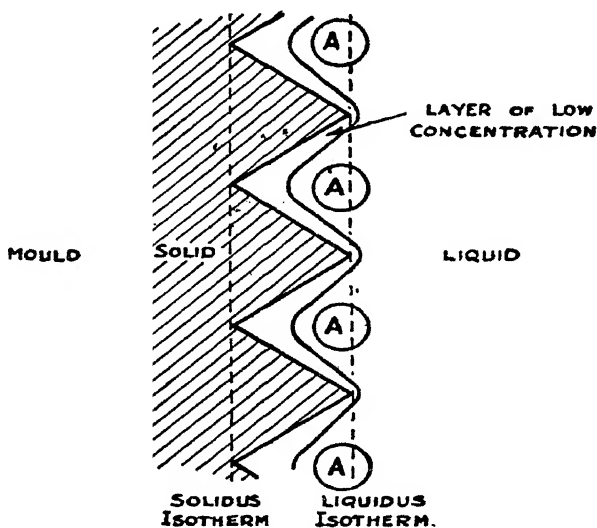


FIG. 39.

Diagrammatic representation of surface of solidification for alloy with a freezing range.

representing liquidus and solidus temperatures coincide. There is no variation in composition of the crystals and therefore no layer of liquid of lower freezing point. Formation of independent centres of crystallisation would thus have to depend entirely on the labile state being reached at some distance from the solid boundary. This would demand an inverse temperature gradient in the liquid, a condition which could not occur. It may therefore be assumed that under the conditions obtaining in an ingot of pure metal (or an alloy having a very small freezing range), solidifying undisturbed in a mould of conductive material, no type of structure other than one solely of

PLATE XXIII.



FIG. 40.
Longitudinal section of "bled"
ingot of 70-30 brass.



FIG. 41.
Longitudinal section of "bled"
ingot of 60-40 brass.

PLATE XXIV.



FIG. 42.

Inner surface of "bled" ingot of 70/30 brass. Actual size.

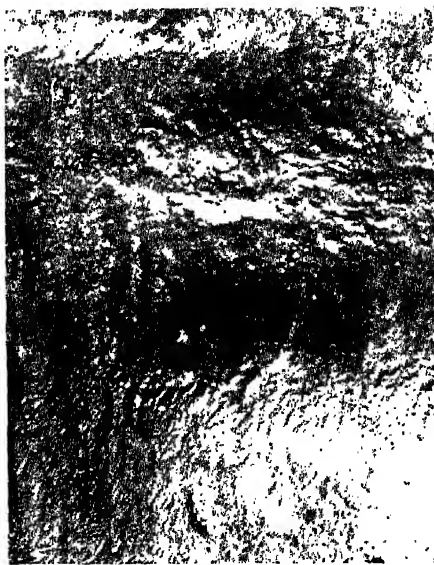


FIG. 43.

Inner surface of "bled" ingot of 60/40 brass. Actual size.

columnar crystals could be obtained. Equiaxial structures, however, are in practice found in such metals, and the inference must be that in these cases the growth of equiaxial crystals has been initiated by inoculation of the (under-cooled) liquid with solid crystal particles. The method to be prescribed for securing under ordinary conditions a largely equiaxial structure in pure metals, or alloys with a very short range of solidification, such as 60/40 brass and 90/10 aluminium-bronze, would depend on the speed of cooling induced by the particular mould used. In the case of a refractory mould it would consist in casting at the melting point of the metal, the necessary solid nuclei being formed during pouring. Where equiaxial structures are found in castings made in a conductive (chill) mould the nuclei have been formed, as has been described, during the turbulence set up by pouring, by the dislodging of portions of the solid surface, and by the washing of splashed metal into the mass. Practical means of obtaining an equiaxial structure, if desired, would thus consist in casting at a low temperature, the rate of pouring being adjusted to give the necessary turbulence concurrently with the extraction of heat by the mould.

The nature of the boundary between the solid and liquid in a solidifying ingot, which has just been discussed, and its relation to the range of freezing of the alloy, can be examined visually in experimental ingots from which the central liquid portion has been removed after partial solidification of the mass. A convenient method of making "bled" strip ingots of brass consists in bottom casting, and striking off the vertical runner immediately after filling the mould in order to allow the still liquid portion of the ingot to drain away rapidly. This method is more satisfactory than the inversion of a top-poured ingot, as, in the "bleeding" operation, both surfaces of the cavity formed are simultaneously freed from liquid. Longitudinal sections of "bled" ingots of 70/30 and 60/40 brass are shown in Figs. 40 and 41, and the surfaces of solidification in Figs. 42 and 43. The ingot of 70/30 brass shows a macrostructure partially equiaxial, and in places bridging of the cavity by projecting dendrites. The upper portion of the surface of the cavity is irregular, with numbers of well-formed octahedra and small sharply pointed dendrites projecting from it. In the ingot of 60/40 brass the structure consists entirely

of columnar crystals, with no bridging of the cavity. The cavity surface is perfectly smooth, both visibly and to the touch. It would appear, therefore, that the explanation given above concerning the way in which independent crystallisation occurs is in agreement with experimental fact as regards the conditions present at the boundary of the liquid and solid phases.

From the practical point of view, the character of the surface of solidification as it progresses from outside to centre of an ingot is important in connection with the occurrence of contraction cavities. The greater the range of solidification of the alloy, the more readily can bridging of the central region, and the formation of cavities, occur in any given form of ingot.

The general features of solidification, and the influence of inherent properties of the material cast, which have been outlined above, indicate clearly those factors which are unfavourable and those which are beneficial in promoting soundness of the ingot. To sum up, it would appear that a parallel mould, a single stream and an alloy solidifying over a range of temperature are factors which tend to increase contraction cavities and entrapped minute gas bubbles, while division of the stream, reduction of its height of fall, adequate feeding and an alloy having a short solidification range operate towards the production of a sound ingot. The direction in which various other factors operate may also be inferred, but the real extent of such influence can only be determined by experiments giving quantitative results. An analysis of the casting process on these lines is described in later chapters.

In certain alloys there is an important connection between the macrostructure of the cast ingot and its behaviour in rolling; this is particularly marked in some types of aluminium alloy and in zinc. The brasses of about 60/40 composition, and possibly other ranges, are occasionally liable to central splitting and opening down the central plane during rolling, but in general brass ingots do not show on cold rolling any pronounced effects due to structure. It would appear probable that the reason for this lies in the relative purity of commercial brass, and in the outer layer of very fine crystals which is characteristic of brass ingots cast by the ordinary method. Where surface cracks are occasionally found they may usually be

traced to "pull" cracks in the ingot surface, caused by penetration of metal into cracks or depressions in worn moulds. This leads to tensile contraction stresses in the "skin" of the ingot, while the metal is passing through the range of mechanical weakness just below the freezing point.

With the development of hot rolling, which introduces conditions likely to increase the influence of small quantities of impurities, the question of effects due to ingot structure is of greater importance, and it is possible that studies on this subject may lead to improvements in manufacture.

Even with cold rolled brass, however, new problems may arise as the result of changes of method, and in the course of research on brass casting several instances of surface breakdown of strip in rolling have been experienced, both in commercial and experimental material. In its characteristic form, the defect consists of fine surface cracks which develop in the early stages of rolling and form folds similar to "spills" on reduction to thin gauge.

In the course of experiments with some pure 70/30 brass ingots cast with an inert mould coating, an initial reduction of 20 per cent. in thickness in one pass before annealing resulted in the development of large numbers of fine cracks on the surface of the annealed strip. That these cracks were caused solely by the rolling treatment is shown by the fact that the 30 ingots of the same batch broken down to half their original thickness in three passes produced no such defects. A further 15 ingots cast from the same materials and reduced by only 30 per cent. in breaking down in three passes showed cracks at the second stage of rolling.

The results of a number of experiments made to investigate the cause of this cracking are included in Table IV. The conclusion drawn is that the lightness of the individual passes given to this material was without ill effect, the cause of the defect being the small total reduction (30 per cent.) given in breaking down. Similar ingots reduced 50 per cent. before the first anneal were free from cracks.

The ingots in question were found to contain 0.02 to 0.03 per cent. lead. Two ingots made from different raw materials, and containing 0.015 per cent. lead did not show cracks at the second stage after an initial breaking down reduction of 30 per

cent. The cracks thus appeared to be fire cracks caused by the presence of 0.02 to 0.03 per cent. of lead, together with internal stress produced by insufficient reduction in breaking down.

TABLE IV
EFFECT OF ROLLING CONDITIONS ON CRACKING OF EXPERIMENTAL 70/30 BRASS INGOTS

Type of Ingot Tested.	No. in Test.	Per cent. reduction in Breaking Down.	No. of Passes in Breaking Down.	Lead Content of Ingots.	Quality of Rolled Strip.
Halves of 12" × 6" × 1" ingot	1 1	50 50	25 3	0.02— 0.03%	Good Good
12" × 7" × 1". Cast from same materials under identical conditions.	31 15 15	50 20 20	3 1 3	" " "	Good All cracked All cracked
12" × 7" × 1" whole ingots.	11 1	40 40	4 4	" "	Good Good
Halves of 12" × 7" × 1" ingots.	4 4 4 4	50 50 50 50	2 3 5 8	" " " "	Good Good Good Good
Halves of 12" × 7" × 1" ingots.	1 1 1 1	30 30 20 20	1 2 1 2	" " " "	Cracked Cracked Cracked Cracked
Halves of 12" × 7" × 1" ingots.	5 5	30 50	3 3	0.024% "	All cracked All good
Halves of 12" × 7" × 1" ingots.	2 2	30 50	3 3	0.015% "	Good Good

Aluminium brass ingots which have a columnar structure are also liable to crack in a similar way, and their behaviour appears to be affected in a like manner by the method of rolling to which the ingots are subjected. Ingots containing 0.2 and 2.0 per cent. aluminium have been found to crack extensively when the breaking down comprised a reduction of 30 to 35 per cent. only, while a reduction of 50 per cent. in breaking down gave distinctly less cracking. The possibility exists that particles

of aluminium oxide were in these cases present in the crystal boundaries and caused fire cracking.

All the ingots referred to possessed a structure consisting largely of columnar crystals, and it is considered that the fire cracking observed was caused by the combination of such a structure and the presence of impurity which segregated in the grain boundaries, causing planes of weakness. The coarser the grain and the straighter the boundaries the more serious the effect. Fuller investigation of the question is needed; such observations as have been made here are necessarily only preliminary and do little more than indicate the potential importance of macrostructure as affecting rolling behaviour. It is, however, clear that defects can be produced on strip independently of the soundness of the cast ingot and there is little doubt that such defects are occasionally a source of difficulty in industrial practice.

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CHAPTER VIII

ANALYSIS OF THE CASTING PROCESS

Casting Temperature, Speed of Pouring and Feeding

Variable Factors—Experimental Methods—Soundness and Structure of Experimental Ingots.

WHILE the knowledge obtained from the examination of a variety of cast and rolled material, the properties of the molten metal and the general features of solidification are essential as a basis for further investigation and valuable as a means of narrowing down fields of inquiry, it is only possible, in the absence of detailed knowledge of the casting process, to lay down working principles of no quantitative significance. Similarly, the mechanism by which certain defects are formed can be explained, but the ordinary method of brass casting, apparently simple on first acquaintance, is actually an extremely complex process in which a number of variables exert a combined influence on the resulting ingot, and such explanations have only a limited practical value. The chief variables are enumerated below, and will be dealt with in suitable groups in the succeeding chapters.

Casting temperature.

Speed of pouring.

Feeding of contraction.

Shape of ingot and position of cooling surfaces.

Thickness of ingot.

Mould temperature and thickness.

Mould coating.

Position of mould during casting.

Gas content of the metal.

The determination of the separate influence of each of these factors requires standardised conditions of working, permitting the variation of one factor while the others remain constant.

The means devised to provide these conditions are described below, and were used principally for the casting of various series of ingots of 70/30 brass.

In view of the probability that the relative influence of the various factors is greatly affected by the size of mould, the dimensions of the standard size mould adopted for experimental laboratory work of the kind described are a matter of considerable importance. The results given by diminutive ingots frequently employed in research on the properties of alloys may be highly misleading when applied to casting problems, and it is worth while to employ moulds which, although smaller than those used commercially, are of sufficient size to render the data obtained applicable in principle to a larger scale. The thickness of the ingots generally used by the authors is 1 inch (similar to that of large-scale ingots), while the width and length, 6 inches and 12 inches respectively, are in the neighbourhood of half those of works ingots. A $12 \times 6 \times 1$ inch ingot of 70/30 brass has a weight of about 22 lb. These relative dimensions are considered more suitable than geometric similarity to the works ingots, since the thickness is the main dimensional factor governing the properties of the ingot, and the effects of varying width and length are more readily allowed for in transferring the data to other types of mould. The intermediate size of ingot adopted would therefore be expected to furnish data showing the direction in which each variable exerts its effect and to lead to conclusions readily confirmed and applied in manufacturing practice.

Cast-iron moulds are used, with walls 1.25 inches thick, smooth-machined internally. In order to minimise effects due to the stream touching the faces of the mould, which is stood vertically, and to avoid variations in rate of flow, a pouring bowl or tundish is used, made by moulding carborundum cement into a triangular shaped trough.* This is held just above the top of the mould by a movable sheet-iron frame. The rate of pouring is governed by the size of the slot or hole in the bottom of the tundish.

Avoidance of accidental defects due to the introduction of

* Moulded carborundum tundishes are strong mechanically and are found to resist wear extremely well.

slag or dross is ensured by the employment of a bottom-pouring ladle consisting of a sheet-iron bowl lined with fire-clay. The nozzle is of carborundum and the stopper of "salamander" rod. A nozzle aperture of 0.5-inch is suitable for the size of mould used. The ladle is held above the mould in a framework, and is internally heated previous to casting. Further advantages of the ladle are the ease of precise regulation of flow and the comparative absence of discomfort to the operator. Loss of heat in the ladle, which is one of the most serious objections to the practical adoption of the method, is much reduced by thickening the lining so that the metal forms a column of height equal to its diameter instead of a shallow pool. With a thin lining a rapid drop from crucible temperature takes place immediately on pouring the metal into the ladle and subsequent fall is slow, the total loss prior to casting possibly reaching 100° C. With thickened walls, previously heated to bright redness by a gas flame, the total loss of temperature need not exceed about 30° C. The apparatus originally used is shown in the photograph, Fig. 44. It is desirable to use copper and zinc of electrolytic quality, with about 25 per cent. of ingot scrap of similar purity. A flux of glass is suitable for melting where standard conditions are necessary.

In the casting operation, the molten brass, at a temperature of 1100° C., is poured direct after rough skimming into the ladle, preheated by gas or by a previous charge. After measurement of the temperature the stopper is raised and the flow regulated to keep the tundish full. The mould is in every case preheated to a temperature slightly above 100° C. and is coated with blacklead or soot to avoid adhesion of the brass. Mould dressings of a type which might interfere with the results by burning or creating gas are undesirable where the effects of other factors are being studied. The surface of the ingots cast without such a mould dressing are exposed to oxidation and thus may be defective compared with commercial material, but in experimental work this provides knowledge of the potential types of defect, which may be partly or wholly changed or avoided by the use of a volatile dressing.

The range of temperature in which it is practicable to cast

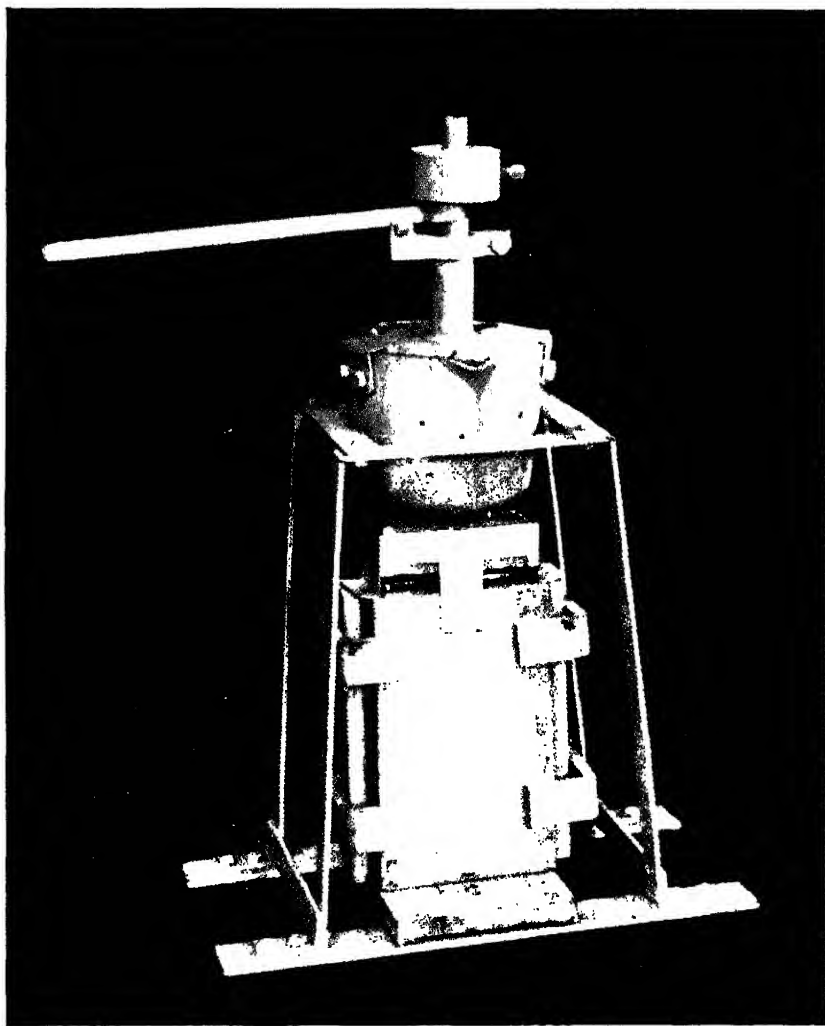


FIG. 44.

Apparatus for experimental casting of 70/30 brass under controlled conditions.

70/30 brass is not more than 200° C. and the range in commercial practice is only about 50 to 100° C. The effect of varying casting temperature is thus adequately shown by a

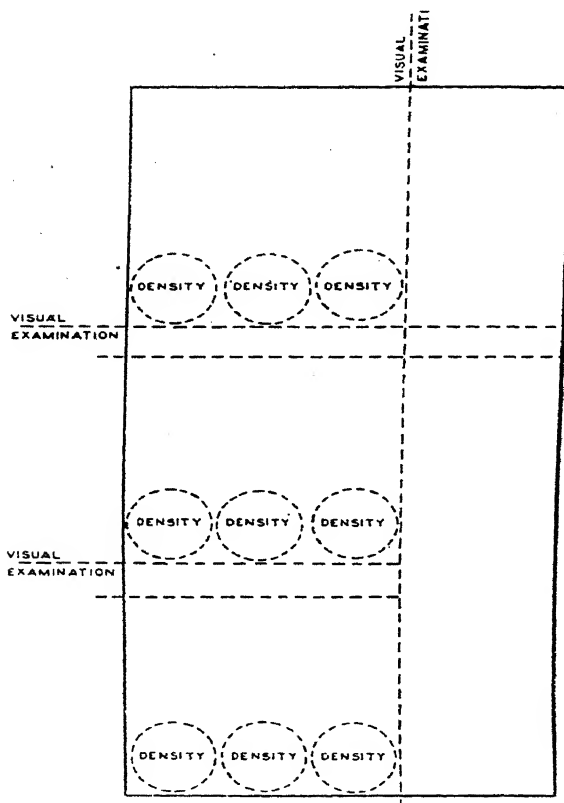


FIG. 45.

Method of sectioning experimental ingots for examination.

series of ingots cast respectively at 1100° C., about the maximum used commercially, at 1060° C., a medium temperature probably slightly below the average temperature used commercially, and at 1000° C., an abnormally low temperature. The variation in rate of pouring which can be conveniently used in a 12×6×1 inch mould ranges from a rate of rise

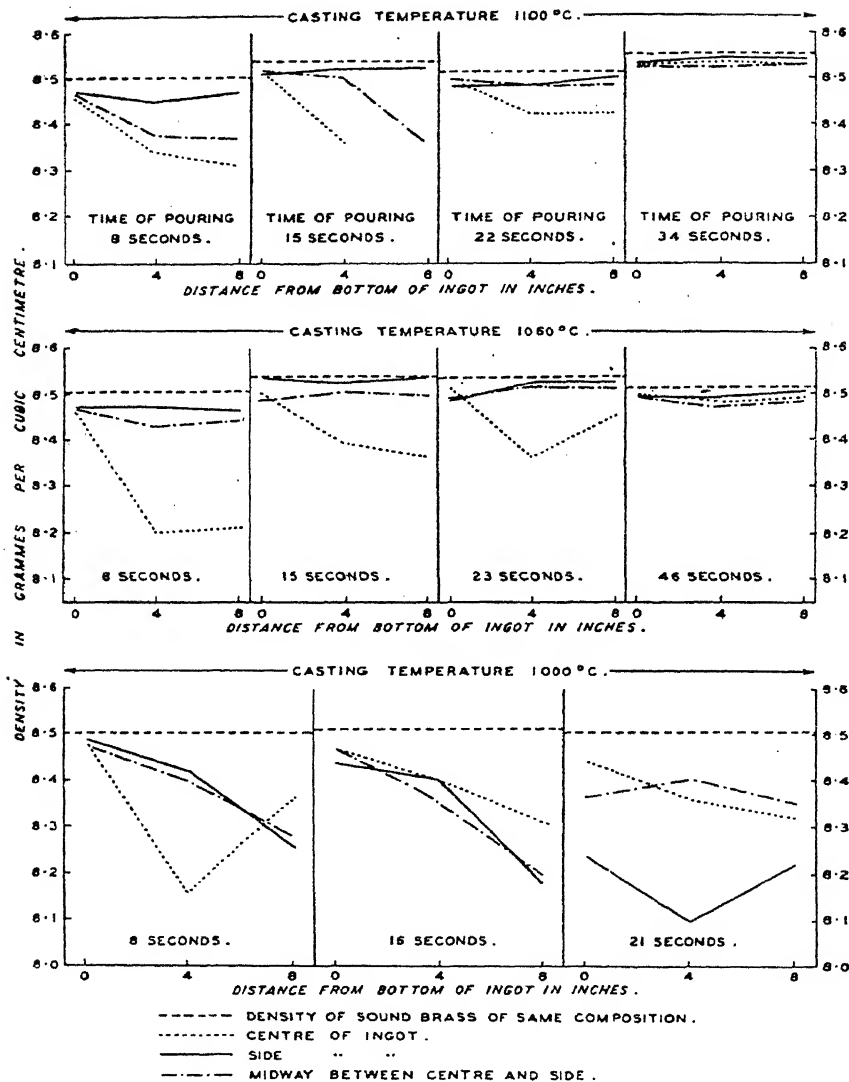


FIG. 46.

Densities of 70/30 brass ingots (unfed) poured at various temperatures and speeds.

of metal in the mould of 1.5 inches per second (about the average speed used in works) to a very slow speed of about 0.3 inch per second.

The method of sectioning and examination is shown to scale in Fig. 45. The effect of variation in casting temperature and speed of pouring on the amount of unsoundness and its distribution in $12 \times 6 \times 1$ inch ingots of 70/30 brass is shown in the density curves Fig. 46 and the composite curve Fig. 47. The values for unsoundness, expressed as percentage volume

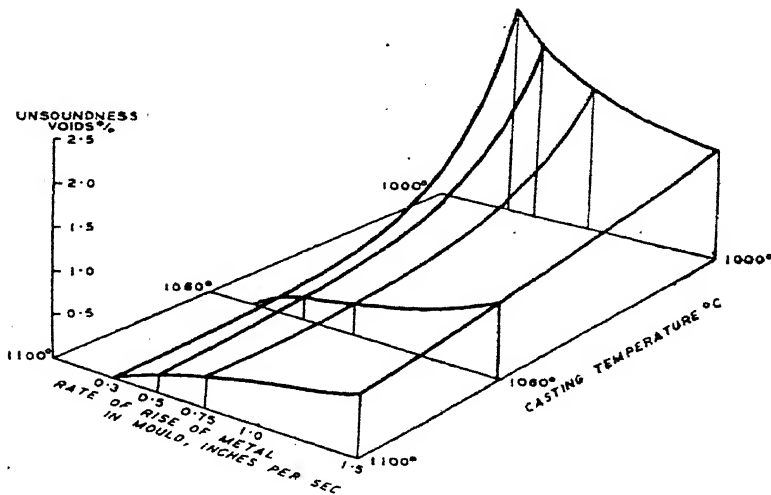


FIG. 47.

Effect of rate of pouring and casting temperature on average internal unsoundness of ingots.

of cavities, do not include the pipe at the top of the ingot. They are calculated from the densities of cylinders cut from each ingot at various positions (see Fig. 45). Sections of certain of the ingots, illustrating visually the influence of variation in casting temperature and speed of pouring are shown in Fig. 48.

The following general conclusions may be drawn from the data given.

With a casting temperature of 1060° C. or above, the unsoundness of the ingot decreases as the rate of pouring decreases. (See Fig. 48 (a) and (b).) This is due to the plane

of complete solidification across the ingot rising closer behind the fluid level with slower rates of pouring, the contraction at the centre being continuously fed by the fluid metal immediately above. A slow rate of pouring of about 0.3 to 0.4 inch rise of metal in the mould per second is necessary to secure internal soundness.*

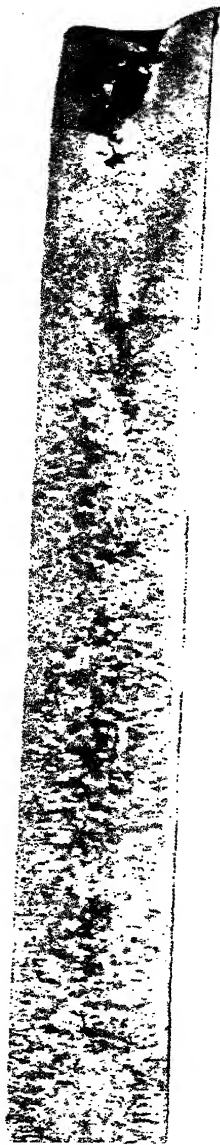
With a very low casting temperature (1000°C.) the ingot is unsound at both rapid and slow rates of pouring, and the unsoundness tends to increase with decrease of pouring rate, owing to the formation of a type of defect differing from contraction cavities and produced by the entrapping of air bubbles by the rapidly solidifying metal. These defects, as shown in Fig. 49, are of a much worse character than contraction defects. They are larger generally, frequently occurring near the surface of the ingot, and on rolling would inevitably lead to "spilly" and badly "laminated" strip, and to the appearance of blisters during subsequent deformation as in spinning.

The solidification of an ingot (of 70/30 brass) of slab form when poured at a high temperature and rapid rate, is such that the central region becomes bridged by crystals growing from the two faces. This region of the ingot is, as a result, inefficient as a passage for molten metal and contraction cavities are formed. With a slower rate of pouring, the contraction cavities are reduced and at the same time the high temperature of the metal allows most of the air injected below the surface to rise and escape. A high casting temperature and slow rate of pouring are thus much more favourable conditions than can be obtained with a low casting temperature at any rate of pouring.

The structural effects are such as would be expected from different combinations of turbulence (pouring speed) and rate of solidification (casting temperature). With a high casting temperature and rapid rate of pouring the penetration of the hot stream produces equiaxial crystals in the central region over a large proportion of the ingot length. These are enclosed by a zone of columnar crystals and an outer thin layer of finely crystalline, rapidly solidified metal. Decrease in pouring speed and the consequently smaller penetration increases the

* A slow rate of pouring is, however, detrimental as regards surface quality.

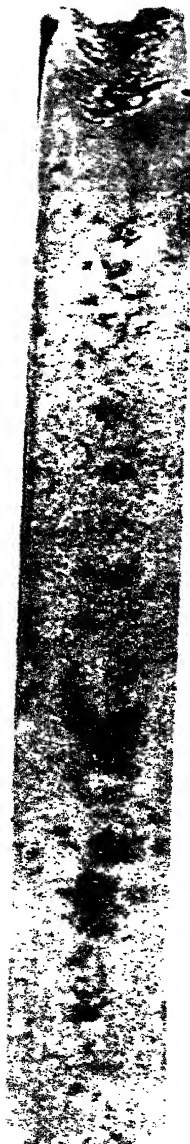
PLATE XXVI.



(a)



(b)



(c)

PLATE XXVII.



FIG. 49.

Cavities due to entrapped air in 70/30 brass ingot poured at low casting temperature. Actual size.

tendency to columnar crystallisation. With a low casting temperature, the existence of solid particles suspended in the liquid within the mould brings about the formation of fine equiaxial crystals.

Good surface quality is only secured by rapid pouring at a high temperature. As either casting temperature or rate of pouring is diminished the surface becomes more defective and under extreme conditions, *i.e.* low casting temperature and slow pouring, defects of abnormal extent are produced if oxidation occurs. These are deep folds and semi-attached particles of splashed metal, and are due to the restriction of free flow by the crust of solid crystals and oxide formed continuously on the exposed liquid surfaces.

The results given, describing the effects produced by variation of casting temperature and speed of pouring, refer to conditions of free natural shrinkage of the ingot and provide a basis for the quantitative expression of the effects produced by additional factors.

The separate effect of feeding is shown by similar data from a series of ingots cast by the standardised methods, with the addition of a feeding operation provided by the use of a "dozzle" or feeding head of fire-clay, held in a recess machined in the top of the mould. In casting, the dozzle is pre-heated and placed in the top of the mould before pouring is commenced. By filling the mould and dozzle without interruption of the pouring, automatic feeding uniform throughout the series is secured.

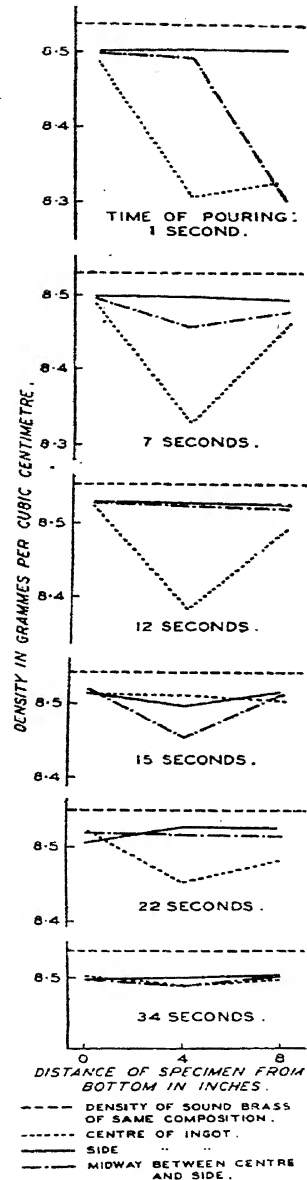


Fig. 50.

Densities of 70/30 brass ingots poured at 1100° C. and various rates. Dozzle-fed.

The results given in Table V and Fig. 50 show the effect of dozzle feeding on ingots cast at a temperature of 1100° C. with the former range of pouring speeds extended to include the maximum possible rate (nearly instantaneous pouring). This is obtainable by the use of a tundish consisting of a crucible slotted through the base, into which the crucible containing the molten brass is inverted. The mould is filled by this method in a time not greater than one second. The data also include values for the density of the whole ingot, determined by weighing the complete ingot, as removed from the mould, in air and water.

While the densities of cylinders machined from different parts of the ingot enable the variations in unsoundness to be mapped out, the density of the complete ingot gives the average for the whole of the material and is the only method suitable for commercial use. In experimental work the results obtained on the whole ingot provide useful supplementary data to those obtained by the method of exploration by small specimens.

TABLE V

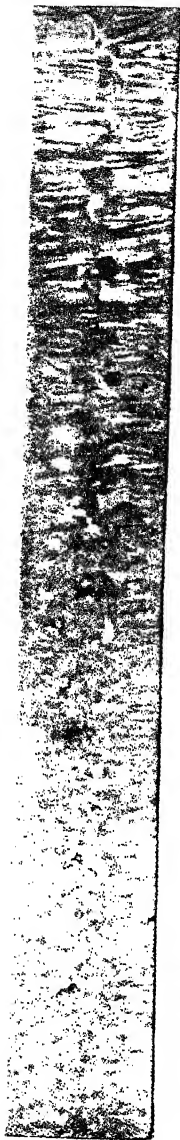
INFLUENCE OF FEEDING ON SOUNDNESS OF 12×6×1 INCH
INGOTS OF 70/30 BRASS POURED AT VARIOUS RATES

Casting Temperature 1100° C.

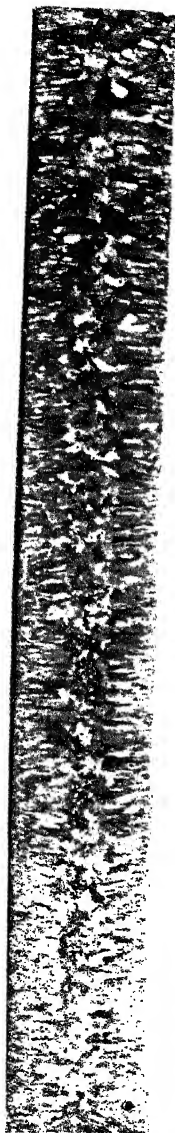
Time of pouring. Seconds.	Copper per cent.	Max. Density Value. Grams per c.c.	Density of whole Ingot. Grams per c.c.	Average Density of Cylinders. Grams per c.c.	Per cent. Volume of Cavities calculated from Density of whole Ingot.	Per cent. Volume of Cavities calculated from Average of Cylinders.	Per cent. Volume of Cavities in unified Ingots similarly cast. Average of Cylinders.
1	70.5	8.515	8.437	8.440	0.92	0.88	—
7	69.6	8.503	8.449	8.456	0.64	0.55	0.93
12	71.3	8.525	8.479	8.505	0.54	0.24	—
15	70.6	8.516	8.466	8.502	0.58	0.17	0.69
22	70.8	8.520	8.483	8.504	0.44	0.18	0.44
34	69.7	8.505	8.435	8.494	0.82	0.13	0.15

The ingot surface is, as would be expected, uninfluenced by the feeding of the top portion.

Etched sections of representative ingots illustrating the effect of feeding are shown in Fig. 51. Central unsoundness, consisting of contraction cavities, is present to a relatively large extent in the rapidly poured ingots, diminishing as the rate of pouring is decreased, and reaching a negligible amount



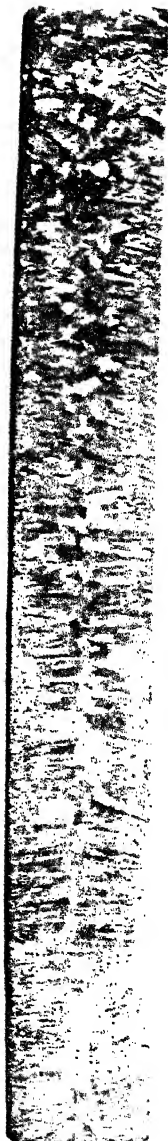
1



12



15



34

Time of Pouring (seconds).

in the slowest poured ingot. The top portion of the ingot, occupied in former cases by the primary pipe, is in all instances, however, quite sound for a length of about 2 inches. The macrostructure of the sections shows as before an increase in columnar crystallisation with decrease in the pouring speed at the high casting temperature used.

Comparison of the last two columns of Table V shows that, whether the ingot is fed or unfed, the extent of unsoundness in the main body of the ingot (below the portion in which piping occurs in unfed ingots) is decreased by decrease in the rate of

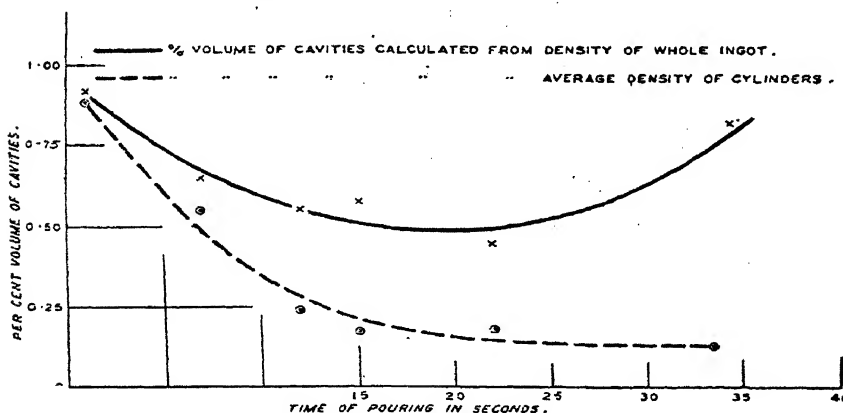


FIG. 52.

Comparison of density of whole ingots and average density of cylinders.
Ingot cast at 1100°C. , exposed to air.

pouring. The average density (as obtained from determinations on a number of cylinders) is, however, raised to some extent by feeding. This is shown by the individual tests to be due to increased soundness of the upper region where, in unfed ingots, there is usually a region of unsoundness below the pipe. The effect becomes less as the rate of pouring is diminished.

The average density obtained by making a determination on the whole ingot, although showing an increase in soundness with diminishing rate of pouring up to a certain point, becomes almost as low in the slowest poured ingot as in the most rapidly poured of the series, and shows no constant agreement with the results obtained from small cylinders (see Fig. 52). The cause of this apparent discrepancy is traceable to the effect of

unsoundness in the surface layers of the ingot which affects the density of the whole ingot but is designedly not included in the machined cylinders, the density of which is a measure of the internal unsoundness only. It will be observed also that the results of the two methods diverge most in the ingot having the worst surface poured at the slowest rate, and approach each other as the surface improves with increasing rate of pouring, practically coinciding in the most rapidly poured ingot. In the last case the surface unsoundness is practically nil and the two methods agree as measurements of internal soundness. A comparison, therefore, of the densities obtained by deter-

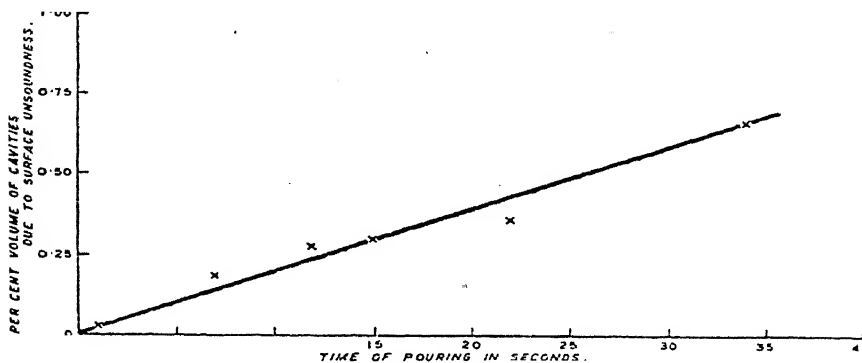


FIG. 53.

Effect of rate of pouring on unsoundness of surface layer of 70/30 brass ingot.
Ingots cast at 1100° C., exposed to air.

minations on the whole ingot and on machined cylinders may be taken as giving a measure of the unsoundness present in the surface layer. Fig. 53, derived from Fig. 52, shows the rate of increase of unsoundness at the surface with decrease of pouring speed, in ingots of 70/30 brass poured in air with no volatile mould dressing.

The series of ingots just described indicates generally that although feeding of the pipe is not capable of filling up completely the central contraction over the whole length of the ingot, the amount of contraction cavities is definitely reduced by feeding. This effect occurs, however, to a variable extent dependent on the conditions of casting. When the rate of pouring is rapid, leaving a large amount of contraction to take

place in the ingot inside the solid shell formed against the mould walls, feeding of the ingot top appears to exert an effect for a considerable distance down the ingot. As a consequence, it is possible to secure a high degree of soundness without employing an excessively slow rate of pouring; in the case of the $12 \times 6 \times 1$ inch mould used in the investigation, the rate of pouring necessary to avoid unsoundness may be raised from 0.3 to 0.8 inch rise per second when feeding is efficiently carried out. The surface quality of the ingot, however, although improved somewhat by the quicker rate of pouring, is still poor. When the rate of pouring is very slow the ingot is internally sound whether fed or unfed.

Under the conditions of dozzle-feeding, the ingot solidifies under a head of fluid metal, the top of which remains practically level while sinking towards the ingot. The effect of dozzle-feeding may therefore be greater than that of the usual method employed, where metal is fed in a stream into the middle of the large head of rapidly solidifying ingot metal. The advantages of the use of dozzle-feeding are :

1. The time spent in feeding is saved, and less care is needed in filling the mould.
2. Discard is reduced.
3. The top portion of the ingot is perfectly sound.
4. The effect on the general soundness of the ingot is probably greater than that of feeding from the crucible.

Against these advantages must be put the slight additional cost, and the additional work of pre-heating and inserting the dozzle.

CHAPTER IX

ANALYSIS OF THE CASTING PROCESS (*continued*)

Dimensions of Ingot. Mould Conditions

Effect of Shape and Size of Ingot—Solidification in Different Forms of Ingot—Position of Cooling Surfaces—Ingot Thickness—Thickness of Mould—Mould Temperature.

THE disadvantage of the parallel form of ingot as regards the difficulty of producing progressive solidification from the bottom upwards by pouring methods which are suitable for avoiding surface defects has already been pointed out. An ingot with parallel or nearly parallel faces is, however, essential for cold rolling, and in general very desirable in the manufacture of strip, in order to avoid rolling difficulties. Nevertheless, it is useful to consider what are the actual effects of change of shape and how far compromise between inconvenience and internal soundness is possible.

Several methods are possible for the production of the necessary conditions for progressive vertical solidification. The method most usually followed, especially in steel manufacture, is the use of ingots increasing in section from bottom to top. In the solidification of such an ingot cast under suitable conditions, an inverted cone of liquid metal is continuously present in the central region and progressively fills the contraction. With the addition of a hot feeding-head or dozzle to avoid deep piping, a sound ingot is produced. The application of this principle to brass slab ingots of the size normally used in manufacture has not been attempted in brass foundries so far as is known. The effect of various degrees of taper on the soundness of 70/30 brass ingots 30 inches in length and 7 inches in width, cast in moulds of wall-thickness 1.5 inches (machined out internally on the lid and back equally to give a straight taper on the thickness of the ingot from bottom to top of 0.5, 0.75 and 1 inch respectively) is illustrated in Fig. 54.

TAPER INGOTS

The data refer to works casting conditions and include a parallel ingot of the same length and width for comparison.

Sections of such taper ingots show few cavities of the type due to entrapped gas, but contain contraction cavities of the normal type; the progressive reduction of unsoundness with increase in degree of taper is distinctly visible in the ingot sections. As in ordinary parallel ingots, the maximum unsoundness is in all cases about the middle of the ingot length. The portions at the top and side of the ingot are sound. The macrostructure shows an increase in the proportion of columnar crystals with increase in the amount of taper. Taper of the ingot has no sensible effect on the surface quality.

It is clear that the maximum amount of taper used in the ingots described, though very large from the point of view of strip manufacture and impracticable as a works method on account of the rolling and manipulation difficulties involved, is only just adequate to reduce effec-

the occurrence of contraction cavities. This is an instance of the difficulty, not unexpected, of applying measures to control the progress of

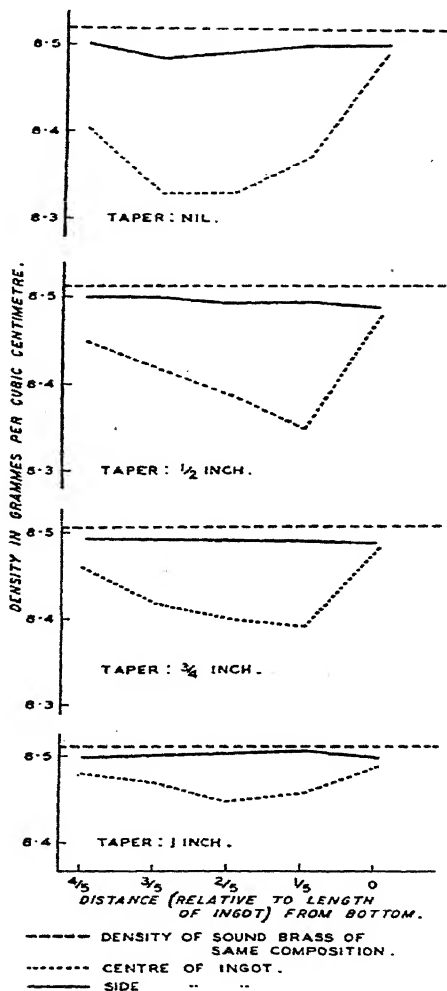


FIG. 54.

Density of 70/30 brass ingots, $30 \times 7 \times 1$ inch, with various amounts of added taper from bottom to top. Casting 1100°C .

solidification under conditions where the time of solidification is extremely short. While an increase of taper above that used in the experiments might give complete freedom from unsoundness, the resulting ingots would be of such abnormal shape as to be useless. It is evident that small modifications in shape of ingot which would be practicable would have no sensible beneficial effect.

While the problem of obtaining progressive solidification from bottom to top of a slab ingot appears incapable of solution, there is the possibility of arranging progressive solidification laterally from side to side of the ingot. This method, though not ideal, would be expected to give increased freedom from unsoundness, and at the same time good surface quality under otherwise normal casting conditions which are not conducive to soundness. A method of graduation of the rate of extraction of heat (chill effect) from side to side of the ingot is not difficult to devise experimentally, and incidentally permits of more means of intensifying such effects than the method of vertical tapering.

The simplest method of inducing the necessary type of temperature gradient consists in pouring with the stream directed close to one side of the mould, which is stood vertically and filled from a tundish according to the standard methods earlier described.

An ingot of 70/30 brass ($12 \times 6 \times 1$ inch) cast by this method possesses a surface of relatively good quality apparently unaffected by the position of the stream. The densities of cylinders cut from different parts show that one vertical half is practically sound and the other half nearest to the stream distinctly unsound (see Fig. 55 (a)). From the examinations of sections it is evident that solidification progresses to some extent directionally from the side of the ingot not disturbed by the stream, and unsoundness is thus concentrated away from the middle of the ingot. Sections illustrating this are shown in Fig. 56 (a). The cooling effect of the mould edge near the stream is, however, considerable and in the ingot described, the main unsoundness, occurring a small distance away from the edge, was distributed over a region of general sponginess. Sections of an ingot in which this cooling effect at the mould edge is decreased by the insertion of a slab of

fire-clay 0.5 inch in thickness (and the same width as the ingot) down the "pouring" edge of the mould are shown in Fig. 56 (b). This clearly indicates that the additional non-conducting layer is efficient in delaying solidification in its neighbourhood until practically the whole of the remainder of the ingot is solid. The density values of the ingot, also illustrated in Fig. 55 (b), indicate that the unsoundness in the ingot is concentrated in the relatively small conical-shaped region near one corner,

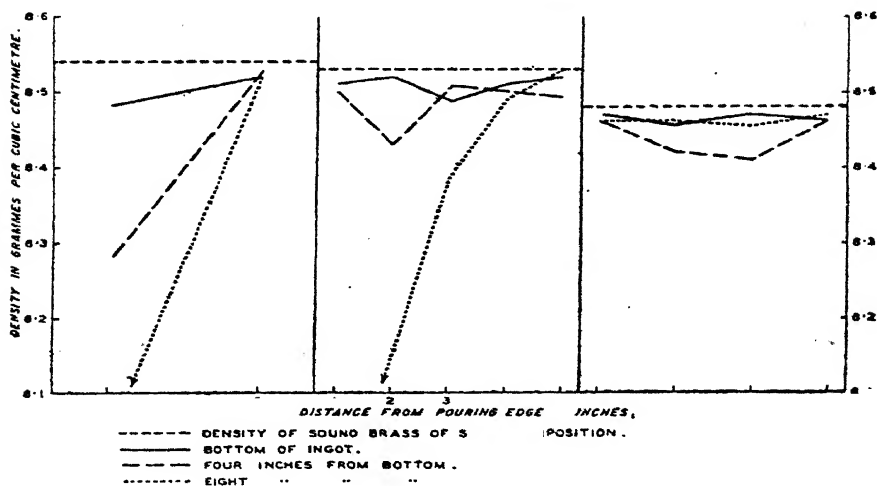


FIG. 55.

Densities of $12 \times 6 \times 1$ inch ingots of 70/30 brass showing effect of directional solidification on soundness. Casting temperature 1100°C .

- a. Stream at one side close to edge.
- b. As in (a) with cooling effect of edge of mould decreased by slab of fire-clay.
- c. As in (b) with addition of dozzle.

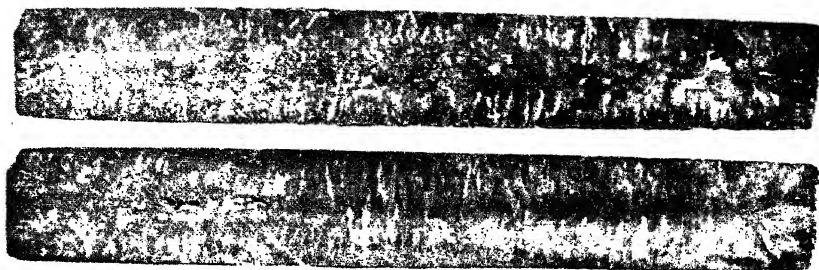
forming a single pipe within 1 inch of the edge of the ingot; the section at 4 inches from the bottom of the ingot (see Fig. 56 (b)) contains only traces of unsoundness at one end. In the unsound region the structure consists of equiaxial crystals, and columnar crystals do not appear within 1 inch from the edge representing the fire-clay slab. It may thus be assumed that the solidification of such an ingot is partly from bottom to top and partly from side to side, having a resultant diagonal direction from the lower cool side of the mould farthest from the stream to the opposite side at the top.

Since almost the whole of the contraction of the ingot can thus be concentrated into a small region, the addition of feeding is likely to be more efficient than under normal conditions. An ingot similar to that just described, but cast with the addition of a dozzle, is shown in Fig. 56 (c). The total amount of detectable unsoundness present is extremely small, little above that present in an ingot poured extremely slowly in the ordinary way; cavities are minute and confined to the central region. The surface of an ingot of this type is similar to that obtained by pouring at the same rate in an ordinary mould. Density values showing the added effect of feeding the localised contraction of the ingot are included in Fig. 55 (c).

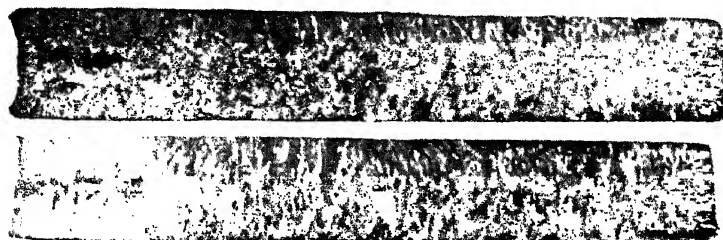
The experimental methods adopted to demonstrate the practical possibility of producing directional solidification in a slab mould with a cooling gradient sufficiently steep to confine all the ultimate contraction to a small region, and the possibility of attaining practically complete soundness by the addition of efficient feeding, may perhaps be difficult to apply in large-scale manufacture. Apart, however, from radical changes in the design and size of ingots used for strip manufacture it would appear that only entirely new measures of this kind are possible in order to secure both good surface quality and freedom from contraction cavities, since in the ordinary method, surface quality is dependent upon rapid pouring while freedom from contraction cavities demands very slow pouring. The principle of the method of directional solidification, therefore, appears sound and presents a wide field for further investigations.

It is generally recognised that soundness is more easily obtained in large ingots than in ingots of small or irregular dimensions. The advantages conferred by ingot thickness are due mainly to the retardation of the cooling effect of the mould faces. The conditions thus provide some approach towards directional solidification from bottom to top of the ingot, and final solidification should occur on the central plane in such circumstances that progressive feeding can take place free from obstacles caused by bridging.

The effect on the character of ingots produced by a large increase in thickness is illustrated by an ingot experiment-



(a)



(b)

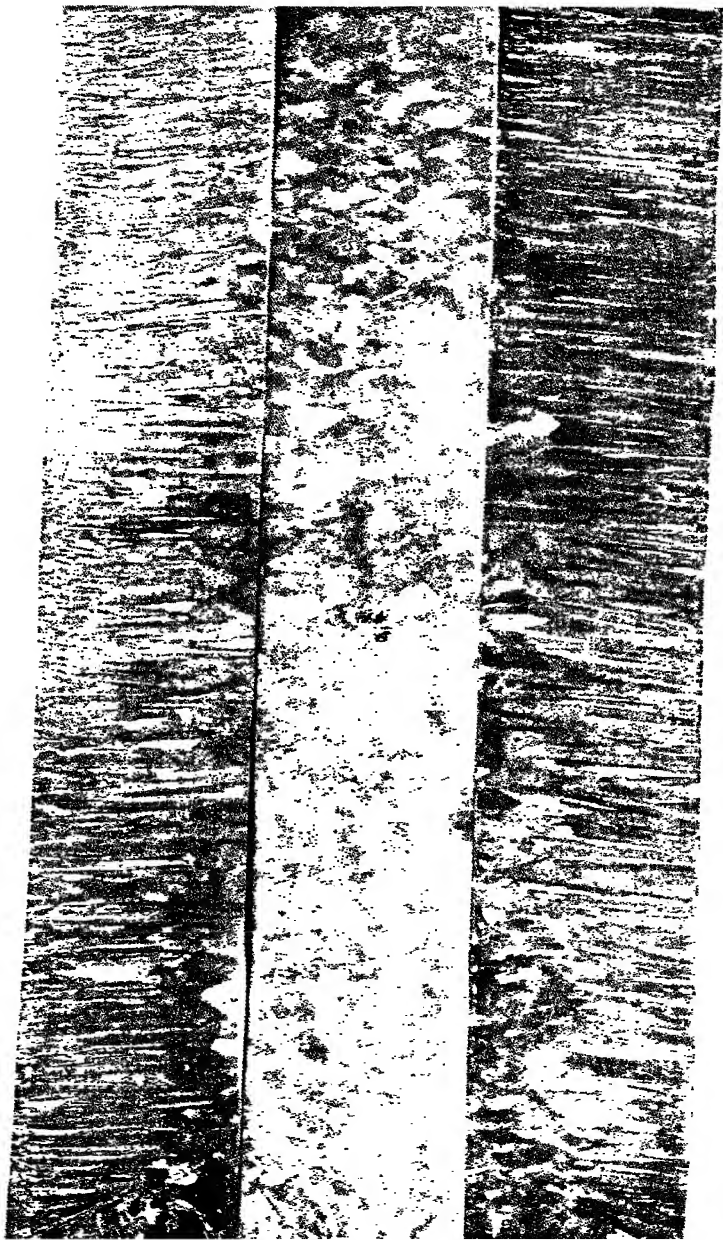


(c)

FIG. 56.

Sections from 12 × 6 × 1 inch ingots of 70-30 brass showing effect of directional solidification on structure and soundness (see Fig. 55).

- (a) Stream at left-hand side near edge.
- (b) As in (a) with cooling effect of edge of mould decreased by slab of fireclay.
- (c) As in (b) with addition of dozzle. $\frac{2}{3}$ actual size.



ally made in a cast-iron mould giving ingots of 24 inches length, 11 inches width and 6 inches thickness, representing a weight of about 550 lbs. This particular thickness is not a suitable dimension for industrial use, but is a large enough increase above the normal thickness to show plainly the effect on the ingot. The walls of the mould were 1.6 inch thick. In casting this ingot the usual foundry procedure was followed, the mould being coated with resin and oil and the metal poured direct from two crucibles simultaneously at a casting temperature of 1130° C. Pouring occupied about 36 seconds and feeding 80 seconds.

An example of an ingot cast in this way and examined by cutting into three slabs parallel to the faces (each slab being 1.75 inch thick when machined) is illustrated in Fig. 57. The sections show little unsoundness due to contraction cavities and no evidence of cavities due to entrapped gas. This is confirmed by determinations of the density at various positions throughout the ingot. These were uniform, showing the absence of concentrated unsoundness, but were generally slightly lower than the value for sound material, indicating the existence of very slight unsoundness throughout the mass of the ingot. As regards internal features generally the ingot may be considered to be of high quality. The macrostructure (see also Fig. 38) was columnar for a depth of 1 inch in the bottom and 2 inches in the upper portion of the ingot, the central zone consisting of coarse equiaxial crystals occasionally separated by small cavities. There was no indication of a central unsound plane and feeding in this type of ingot is evidently almost complete.

Examination of the strip produced by the rolling of the cast slabs showed that while a few typical spill defects appeared on the surface corresponding with the original ingot surface, only very small defects occurred on the strip surface corresponding with the cuts through the ingot interior. Spinning tests and microscopic examination of the strip showed almost complete absence of subsurface defects.

The features described indicate that by increasing the thickness of the strip ingot to about four times the dimension commonly used, conditions are obtained which give a near approach to soundness throughout the mass. This is due

mainly to reduction of the cooling effect of the mould faces, and emphasises the view that the chief factor militating against the production of sound metal in the ordinary slab ingots is the rapidity of solidification across the ingot section. The manufacture of cold-rolled strip from the thick type of ingot experimentally used would undoubtedly be uneconomical for works conditions, but the useful generalisation may be regarded as confirmed that an increase in ingot thickness wherever possible is a desirable step from the point of view of quality of the final product.

Apart from variation in size and shape of the cast ingot itself, a knowledge of the principal effects of variation in mould wall thickness on the rate of solidification of the thin slab type of ingot is of considerable importance from the practical point of view, since if the effect were large, it would be predicted that in a slab mould with walls of varying thickness, ingots would be produced with the central region of cavities deflected towards the thinner wall, with the consequent danger of the formation of surface and subsurface defects during working. The question also has some connection with the design of moulds.

Measurements of the rate of cooling through the temperature range 900° to 600° C., of ingots of 70/30 brass cast by the experimental standardised methods, in $12 \times 6 \times 1$ inch cast-iron moulds of various thicknesses, show no great variation when the wall thickness of the mould is reduced from 1.25 inch to 0.4 inch. Similar measurements made in moulds lined with Acheson graphite 0.5 inch thick or with carborundum plates give considerably slower rates of cooling. The lack of marked influence of the variation in thickness of cast-iron moulds on the rate of solidification is reflected in the small variation in unsoundness of the ingots, as shown by the values given in Table VI.

The structure of the brass is similarly unaffected. It is clear that the disturbing effect of the stream of metal does not mask the effect of mould thickness to any appreciable extent, since ingots "instantaneously" cast in the same moulds (time of pouring about 1 second) show the same absence of notable variation in soundness and structure.

It is evident that within the limits of these experiments the rate of solidification of brass in contact with the cast-iron

Mould 1.25 inch thick cast iron.



(a)

Mould 0.5 inch thick graphite.

Mould 1.25 inch thick cast iron, uncoated.



(b)

Mould 1.25 inch thick cast iron, lamp-black coating.

Mould 0.25 inch thick cast iron.



(c)

Mould 1.25 inch thick cast iron.

FIG. 58.

Ingots cast in moulds having opposite faces of different characters, showing negligible effect of thickness of cast iron and of mould coating on rate of solidification.

TABLE VI

INFLUENCE OF MOULD THICKNESS ON SOUNDNESS OF
 12×6×1 INCH INGOTS OF 70/30 BRASS
 Casting Temperature 1100° C. Ingots dozzle-fed.

Thickness of Mould Wall.	Ratio Cross-sectional Area.	Volume of Cavities per cent.
<i>Time of pouring</i> 10 seconds.	Mould : Ingot.	
1.25 inches	4 : 1	0.44
0.75 inch	2.3 : 1	0.65
0.5 "	1.3 : 1	0.59
0.25 "	0.7 : 1	0.76
<i>Time of pouring</i> 1 second.		
1.25 inches	4 : 1	0.92
0.75 inch	2.3 : 1	0.97
0.5 "	1.3 : 1	0.76
0.25 "	0.7 : 1	1.39

surface is nearly independent of the thickness of the cast iron. This is not in agreement with the view which has been widely held by foundrymen and metallurgists. Confirmation is, however, readily obtainable by a comparison in the same ingot of the rates of solidification produced by widely varying thicknesses of cast iron. When ingots are cast in a composite mould, the back face being of cast iron 1.25 inch thick, and the lid being either cast iron 1.25 inch thick with the surface heavily smoked, cast iron 0.25 inch thick, steel 1.25 inch thick or Acheson graphite 0.5 inch thick with steel backing, any difference in the rate of solidification from the two faces of the mould should be revealed by a displacement of the central contraction plane towards the face having the smaller cooling influence. Actually a definite effect was only observable in the ingot poured in the mould having one face of Acheson graphite, the contraction plane being in this case about 0.1 inch nearer the graphite face than the central line. Fig. 58 shows a series of longitudinal sections of ingots cast in composite moulds, illustrating the features described.

It may be definitely concluded that as far as thin slab ingots of brass are concerned, variation in thickness of the cast-iron mould within practicable limits as regards mechanical strength

is negligible in its effect on the ingot. The solidification of the ingot is governed primarily by the heat capacity of the mould, and from this point of view, only a small thickness of cast iron is required. For brass slab ingots of about 1 inch thickness the minimum thickness of mould giving approximately the same chill effect as any greater thickness, appears to be about 0.25 inch, and any increase in thickness beyond this minimum may be regarded almost entirely from the point of view of mechanical life of the mould. In the design of cast-iron moulds, therefore, the question of chill may be entirely disregarded.

The temperature of the mould as prepared before use in normal brass-casting practice is about 100° C., mainly in order to avoid any moisture on the inner surface. From certain points of view a higher mould temperature would have advantages, particularly in facilitating the remelting of splashed particles of metal and so giving improvement in surface quality of the ingot. It is found in practice, however, that high mould temperatures are a distinct disadvantage because of other types of surface defect which are introduced.

Experimental ingots poured in cast-iron moulds heated to initial temperatures varying from 36° C. to 400° C. show that up to 200° C. the ingot surface is improved slightly, but above this temperature there is definite deterioration. The defects given by high mould temperature are holes of varying depth, sometimes extending a considerable distance into the ingot. In general the holes have sharp edges, and appear to have been produced by the propulsion of gas into the ingot surface during the period of solidification. This phenomenon, known as "blowing," may be met with on the back surface of commercial crucible-poured ingots where the impingement of the stream of metal during pouring has resulted in a local overheating of the mould face. The cause is dealt with in detail in Chapter XII dealing with mould materials.

In order to determine the separate effect of mould temperature above 200° C., in the absence of the factors giving rise to the blowing defect, it is necessary to avoid cast iron and to use moulds of a material such as mild steel. Density measurements given in Table VII indicate that variations of mould temperature within the limits likely to occur in practice would not affect appreciably the soundness of the ingot. The

structure is also similarly unaffected. For cast-iron moulds such as are used in commercial brass casting, the mould temperature of about 100° C. normally used is satisfactory for general purposes.

TABLE VII

INFLUENCE OF MOULD TEMPERATURE ON SOUNDNESS OF
12×6×1 INCH INGOTS OF 70/30 BRASS

Casting Temperature 1100° C. Ingots not fed.

Mould Temperature °C.	Percentage volume of Cavities.	
	Ingot poured in 10 seconds.	Ingot poured in 30 seconds.
36	2.77	1.94
95	2.84	1.26
190	2.93	1.91
400 steel mould	2.69	1.47

The effect of further increase in mould temperature above 400° C. on the rate of solidification of brass, determined on

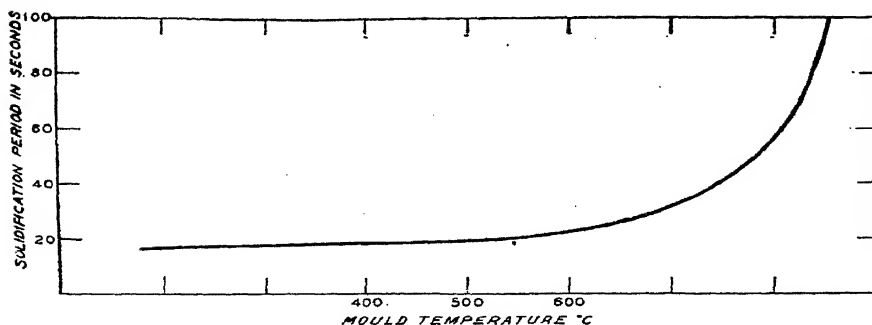


FIG. 59.

Variation of solidification period with mould temperature in 70/30 brass ingots 1.5 inches in diameter.

small cylindrical ingots cast in steel moulds heated to temperatures varying from 100° to 850° C., is shown in Fig. 59. This determination of the limiting mould temperature beyond which a considerable delay in solidification occurs is of interest,

mouth of the mould; these conditions have been shown in Fig. 5, which illustrates the ordinary method of crucible casting (see Chapter II). The flame gases are given off with such rapidity during pouring, that particles of molten metal can commonly be seen being projected out of the mould. No definite knowledge of the complete functions of the mould coating and of the conditions which demand its use or determine the suitability of the various mixtures, has hitherto, however, been available. Various suggested explanations have been put forward speculatively from time to time; these include such functions as lubrication, protection of the mould, thermal insulation, prevention of oxidation, and sealing of the porous surfaces of the mould.

In the light of the knowledge obtained by the examination of strip and ingots, particularly as regards the types of ingot defect which develop and become surface defects on rolling to strip, it would be suspected that, although the inflammable type of mould coating fulfils useful functions, it might be a contributory factor in the production of subsurface ingot cavities. Cavities of this type are not generally formed in a mould coated only with refractory material such as blacklead, clay, etc., except by the use of an extremely low casting temperature. These casting conditions do not occur frequently in industrial practice, and the only alternative appears to be to connect the common type of subsurface cavity with mould surface conditions (including the mould coating). On these grounds it is clear that, as a detail of the casting operation and as a factor probably influencing the characteristics of the ingot to a considerable extent, the study of mould coatings is a subject of major importance.

Direct observation of the behaviour of mould coatings during ingot casting is difficult and their investigation can be carried out more suitably by special experiments on small specimens of brass, melted and solidified under controlled conditions simulating those present at the mould surface during the casting of ingots.

The small-scale method devised by the authors consists in the pouring of a definite volume of brass under standard conditions on to a horizontal slab representing the mould. The apparatus is shown in Fig. 60. A wire-wound electric furnace,

containing an enclosed carborundum crucible arranged for bottom pouring, is supported at a fixed height above a base

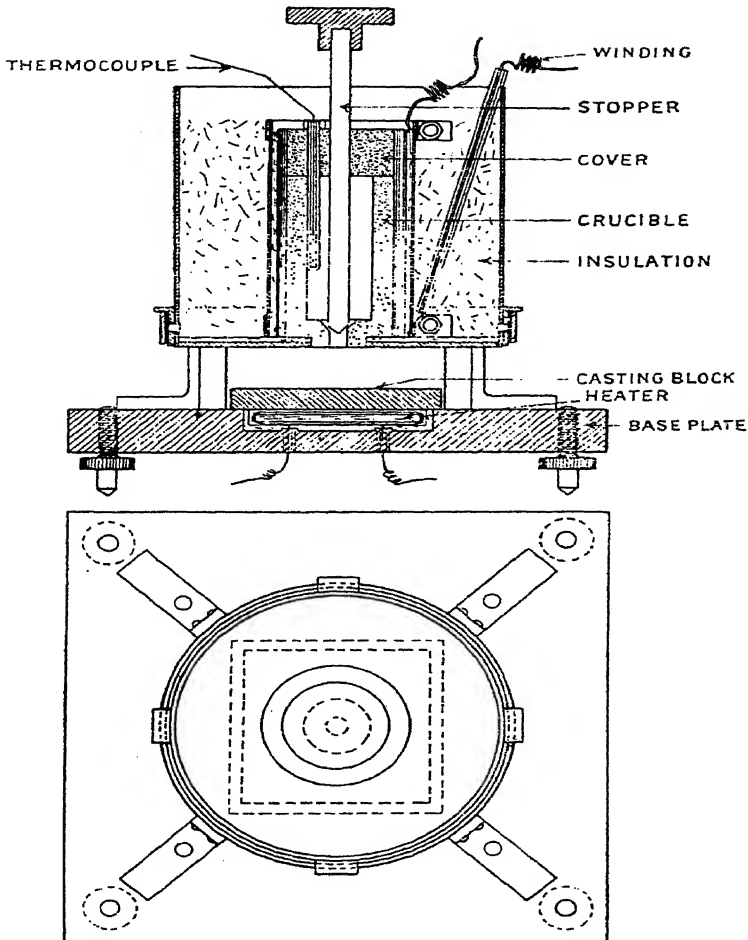


FIG. 60.

Experimental apparatus for investigation of behaviour of mould coatings.

on which is placed a plate of the mould material to be tested. The plate of mould material is heated to the required temperature by a wire-wound heater in a recess in the base. Thermo-

couples of platinum-platinum-rhodium are provided for controlling the temperature of the crucible and that of the mould material. The apparatus is fitted with levelling screws. A machined block of 70/30 brass, 180 grams in weight, slotted to fit around the stopper of the bottom-pouring crucible is inserted, melted and raised to the required casting temperature (generally $1050^{\circ}\text{C}.$). On lifting the stopper the molten brass is made to fall in a smooth stream on to the surface of the plate of mould material, solidifying as a disc about 3 inches in diameter.* Examination of discs cast in this way under exactly reproducible conditions can give a considerable amount of information concerning the behaviour of both the brass and the mould material. At the centre of the disc the continuous impingement of the stream on the slab of mould material gives severe conditions, simulating those in the neighbourhood of the stream during the casting of an ingot. Towards the edge the amount of flow of the brass over the mould material is small and the conditions probably approximate to those in the regions of an ingot away from the stream. Measurements of the thickness of the solid disc are made to determine any variations due to different states of oxidation of the alloy, the standard for comparison being a similar disc melted and solidified in an atmosphere of hydrogen.

The details of some experiments and the thickness and appearance of the discs are given in Table VIII.

Discs cast on a clean machined cast-iron slab show an upper surface greyish-yellow in colour covered with relatively deep ridges produced by the presence of a surface layer of oxide on the flowing metal during casting. The thickness of the discs is 0.75 cm. The under-surface shows less oxidation, and is generally smooth except near the centre, where deep pits

* An approximate estimation of the surface tension of a liquid metal can be obtained by measurement of the thickness of the free edge of the cast slab of approximately circular shape, poured on a horizontal surface, provided free static conditions are reached before solidification begins. Assuming the metal does not wet the plate the surface tension increases with thickness according to the formula :

where T is the surface tension in dynes per cm., g is 981 (the gravitational acceleration in cm. \div sec.²), d the density of the metal and h the thickness of the disc.

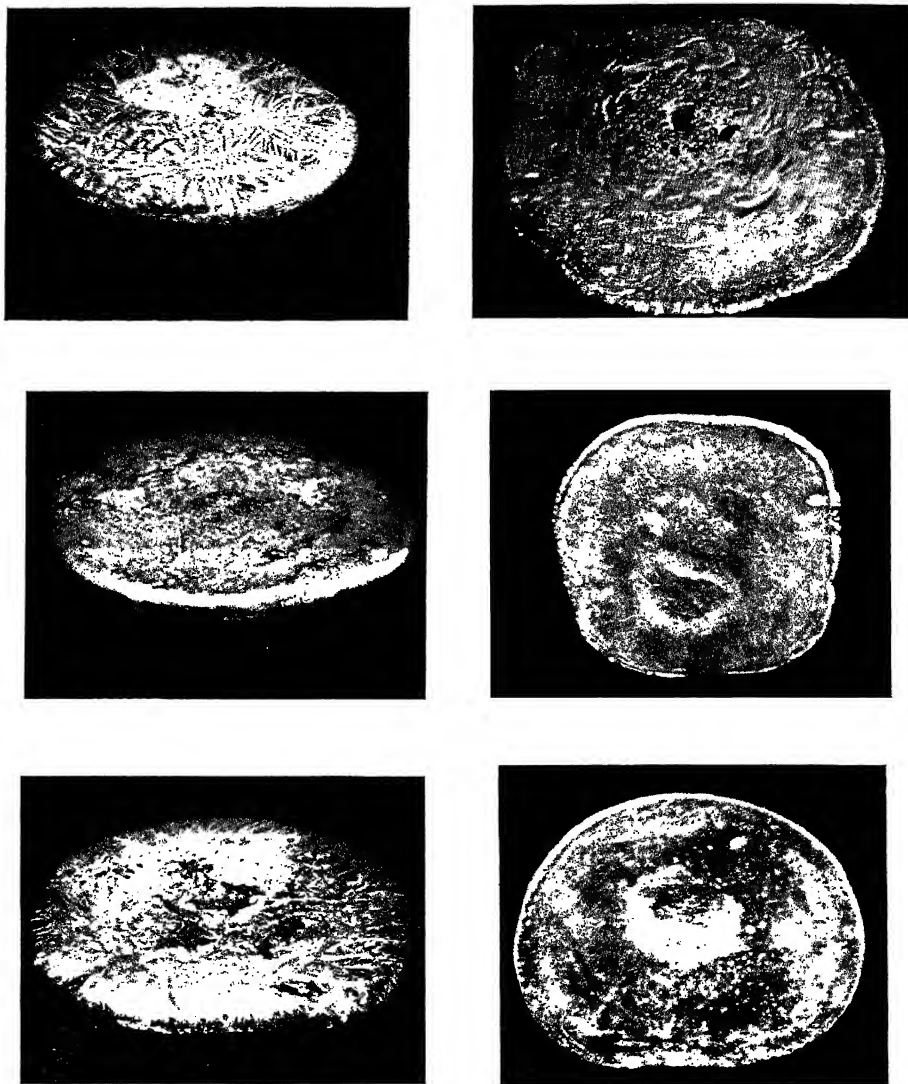


FIG. 61

Small brass discs poured on horizontal cast-iron plate coated with different materials.

- (a) No coating; "blowing" at point of impingement, and oxide skin on free surface.
- (b) Coating lard oil; smooth under surface and comparative freedom from oxidation.
- (c) Coating lamp-black, with reducing coal-gas atmosphere; smooth surface and freedom from oxidation.

occur, similar in appearance to blowing. The appearance of a typical disc made in this way is illustrated in Fig. 61 (a).

TABLE VIII

CAST DISCS OF 70/30 BRASS. INFLUENCE OF MOULD-DRESSING
ON THICKNESS AND SURFACE
DISCS POURED IN AIR ON HORIZONTAL CAST-IRON PLATE
Temperature of Plate 100° C.

Dressing.	Thickness of Disc in cms.	Nature of Under-surface of Brass Disc.
None	0.745	Wrinkled and deeply pitted.
Lamp-black	0.692	Very smooth.
Graphite	0.706	General roughening.
French Chalk	0.710	Slightly rough.
China Clay	0.717	Very deeply pitted.
Fire-clay	0.695	Deeply pitted.
Alumina	0.736	Slightly rough.
Zinc Chloride	0.603	Slightly rough.
Fused Borax	0.666	Included globules of borax.
Aluminium Powder	0.734	Deeply pitted. No reaction with aluminium.
Magnesium Powder	0.662	Slight general roughening. No reaction with magnesium.
Lard Oil (thick)	0.594	Very smooth.
Resin	0.544	Very smooth.
Resin and Lard Oil	0.604	Very smooth.
Graphite and Lard Oil	0.615	Very smooth.
Paraffin Wax	0.628	Smooth wavy surface.

Coatings of various non-volatile dressings, such as lamp-black, graphite, chalk and metal oxides, applied to the cast-iron plate before casting, produce no variation in the surface characteristics of the disc and little effect on the thickness. These coatings have the effect, however, of reducing or avoiding altogether the blowing defects on the under-surface of the disc. Zinc chloride, a material of high volatility (boiling point 730° C.) gives a considerable reduction in thickness to about 0.60 cm., with a relatively smooth surface. This effect appears to be due to reaction between the zinc chloride and the oxide on the brass, with the result that flow of the liquid metal is not impeded. Metallic coatings of aluminium and magnesium appear to remain unaffected by the molten brass and exert no visible effect on the characteristics of the cast disc.

With coatings of volatile materials, such as various oils, resin, and mixtures containing these, a markedly different type of disc is produced. Immediately on commencement of pouring the coating begins to volatilise, giving off a cloud of flame and smoke which completely fills the space under the furnace. The resulting disc is generally bright except for carbon stains, and is always considerably thinner than those cast with non-volatile coatings. With a heavy coating of resin, for instance, the disc obtained is only about 0.55 cm. thick. The appearance of a typical disc (cast with lard oil) is shown in Fig. 61 (b).

It is evident that the absence of oxide on the disc is due to isolation of the brass from the air by the cloud of gas evolved from the volatile coating, but in view of the violence with which the gases are given off it would appear unlikely that the increased fluidity is wholly due to the avoidance of oxidation of the brass. This is confirmed by comparison with a disc made by melting a block of brass in an electric furnace on a carbondum slab in hydrogen gas. The thickness of the resulting disc is 0.69 cm., *i.e.*, less than that of one cast in air from the disc-casting furnace, but appreciably greater than the thickness of discs made with a volatile mould coating. The surface of a disc solidified in hydrogen is brightly metallic, showing no trace of tarnish. If similarly cast in an atmosphere of cylinder nitrogen the disc shows oxidation due to a small percentage of oxygen present in the nitrogen, and the thickness is about 0.70 cm. Discs made in the same apparatus with free admission of air are heavily coated with oxide and are considerably thicker, about 0.85 cm.

The effect of the presence of surface oxidation on the apparent fluidity of molten brass may thus be considerable. It would also be inferred that the impetus of the stream in the casting of discs as first described produces a spread of the metal which results in a reduction in thickness of the disc. A part of the apparent increase in fluidity of the discs cast with volatile coatings is evidently due to the effect of the evolved gases in reducing oxidation and the remainder to the mechanical propulsive effect of the gas evolution on the flowing brass.

These results, viewed in the light of observations on the practical casting of brass ingots in iron moulds, suggest an

explanation of the good surface quality of ingots cast by the usual methods, which include the employment of an inflammable mould-dressing.

It would appear that this type of mould-dressing has several separate functions, the most important being due respectively to its volatilisation (and partial combustion) during casting and to the residual coating of carbonaceous matter left in the mould. A strongly reducing atmosphere formed within the mould and round the stream of molten brass acts as a shield preventing contact with the air. Without this effect oxidation would occur both on the surface of the stream and on the surface of the metal rising in the mould, producing an adherent oxide skin. Since the stream must inevitably produce waves and splashes during the filling of the mould, the ultimate effect on the final ingot surface would be the formation of folds, and other defects due to globules of metal not remelted into the mass of the ingot.

The second function of the mould-dressing is the protection of the mould surface from undue heating by the molten metal, so avoiding the projection of gases from the cast iron into the surface of the ingot. This protection is afforded by many types of material and is not peculiar to the inflammable mould-dressing.

These conclusions are confirmed by experimental discs cast on plates of mould material coated with an inert dressing such as carbon (soot), with a reducing atmosphere applied externally. With the conditions of pouring previously described, and the use of a large coal-gas flame to fill the space between the furnace and the plate of mould material, the resulting disc is about 0.65 cm. in thickness. This is in agreement with the former data when allowance is made for the absence of turbulence due to gases from the coating. Similarly, the use of a sheet-iron box, fitting closely to the base of the furnace, enclosing the plate of mould material and maintained full of coal-gas which burns from a hole in the top of the box round the stream of metal, gives a disc of about 0.67 cm. in thickness. Such discs are unoxidised, showing a smooth surface similar to those given by the use of a volatile mould coating. A typical disc cast with a coal-gas flame is shown in Fig. 61 (c).

Discs poured on a plate of cast iron in the clean machined

condition, using a coal-gas flame, show the central blowing defect and also some general irregularity of the under-surface. This suggests that the mould coating, apart from its influence in avoiding "blowing," has the effect of reducing the rate of extraction of heat (chill) when the molten metal initially comes into contact with the mould surface. A further function of the dressing applied to ingot moulds is thus the provision of a small degree of heat insulation. In this connection it may be noted, on examination of sections of ingots cast in a mould coated only on one face (see Fig. 58 (b)), that the central contraction cavities are not displaced towards the coated face. It would appear, therefore, that the degree of heat insulation provided by a mould coating affects the surface without influencing appreciably the solidification of the mass.

It has been common practice to cast brass of about 60/40 composition, used for hot rolling, in open shallow moulds. When cast in this way, one face of the slab ingot solidifies in contact with air, and to avoid oxidation it is usual to sprinkle powdered borax over the surface to form a protective slag, after raking off the oxide and other foreign matter accumulated during the casting operation. A non-inflammable refractory mould-dressing is used. The ingots show surface defects, but these are normally not sufficiently extensive to persist throughout the rolling and reheating processes. This casting method, in view of the extensive exposure to air which occurs, is of interest in connection with the use of methods of avoiding oxidation.

Ingots of 70/30 brass cast in an open half of a $12 \times 6 \times 1$ inch mould by pouring from the crucible into one corner show an extremely rough grey-black upper surface. During casting the successive waves of molten metal pile up the previously oxidised surface into a series of ridges. The under-surface also shows smaller defects of the same character. These are illustrated in Fig 62 (a) and (b). In an ingot poured in the same way but with a flame of coal-gas covering the mould, no ridges are found, and the resulting surface is smooth and not discoloured. No appreciable oxidation occurs and the only defects are fissures due to contraction which causes sinking of the inter-dendritic liquid in the final stages of solidification. In these regions the surface dendrites are sharply defined in a



(a)



(b)

FIG. 62.

Surfaces of 70/30 brass ingot cast in horizontal open mould freely exposed to air during pouring, showing serious oxidation defects.

- (a) Top surface.
- (b) Under surface.



(a)



FIG. 63.

Surfaces of 70/30 brass ingot cast in horizontal open mould in reducing atmosphere of coal gas, showing contraction fissures but absence of oxidation defects.

- (a) Top surface.
- (b) Under surface.

similar manner, but not so clearly as those found on cast slabs of tin and antimony. The appearance of such an ingot is shown in Fig. 63 (a) and (b). Comparison of the two ingots demonstrates fully the influence of oxidation on the behaviour of molten brass during casting. Similar ingots of 60/40 brass, made using the protective gas flame, are free from interdendritic fissures, and the surface of the ingot is smooth, showing only shallow shrinkage depressions not more than 0.1 inch in depth. The difference is due to the small range of temperature over which the alloy freezes.

Such experiments, besides illustrating the effects of oxidation, incidentally indicate an alternative method for the casting of open-mould ingots of 60/40 brass. They show also that the method is not suitable for 70/30 brass or other alloys freezing through a range of temperature of more than about 10° C.

The results given by the casting of small discs are confirmed by extending the experiments to the casting of $12 \times 6 \times 1$ inch ingots in a cast-iron mould, the conditions of casting being varied to compare with those used on the smaller scale. Ingots cast in air without a mould-dressing or with dressings of non-inflammable character show an irregular ridged surface with much subsurface inclusion of oxide films; small globules of metal, produced originally by splashing during pouring, fall from the ingot on removal from the mould and other similar particles can be easily detached from the surface. By using an inflammable mould-dressing a smooth surface is obtained, the only defects being small pits and adherent particles of carbonised dressing. With a low casting temperature, however, defects caused by incompletely remelted splashed metal can be produced. By omitting the dressing and substituting a copious coal-gas flame enveloping the top of the mould and the stream of metal, ingots are produced free from surface oxide, but showing "blowing" as well as folds due to the rapid initial chill effect of the bare mould surface. These defects are eliminated by applying an inert insulating dressing, such as soot from an acetylene gas flame or a wash of refractory material, and avoiding direct impingement of the stream on the mould face. Similarly, smooth surfaces are produced by closing the mould with a well-fitting iron top provided with inlet and outlet pipes for gas, and pouring into a mould filled

with an atmosphere of coal gas or hydrogen. The surface qualities of typical ingots cast with various mould-dressings and in different atmospheres are described in Table IX.

TABLE IX

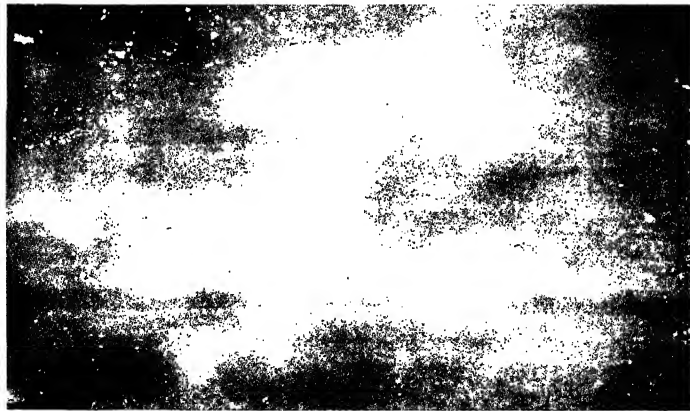
CAST INGOTS ($12 \times 6 \times 1$ INCH) OF 70/30 BRASS
INFLUENCE OF MOULD-DRESSING AND ATMOSPHERE ON INGOT SURFACE

Casting Temperature 1100°C . Time of pouring 10 seconds.

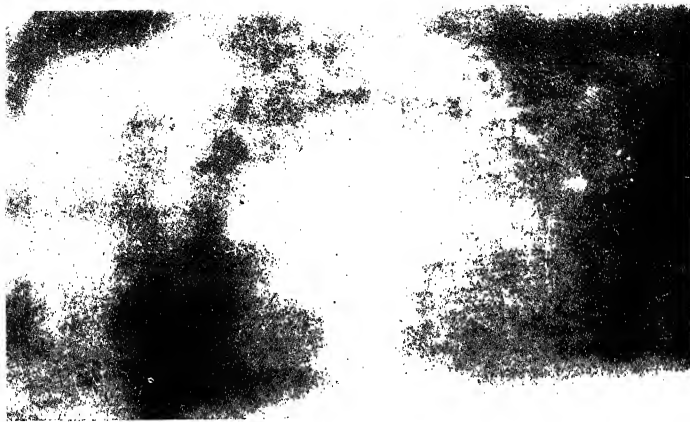
Atmosphere.	Mould-Dressing.	Remarks on Ingot Surface.
Air	None	Much oxide skin. Few unmelted splashes.
"	Resin	Surface free from oxide skin but pitted. Few splashes.
"	Lard oil	Free from oxide skin. Many splashes and pits.
"	Lard oil and graphite	Free from oxide skin. Small number of pits.
"	Paraffin wax	Little oxide skin. Many splashes.
"	Turpentine	Little oxide skin. Few pits and splashes.
"	Zinc Chloride	One face covered with oxide skin. Very large number of splashes.
Carbon dioxide	None	Much oxide skin. Some splashes and pits.
Nitrogen	"	Similar to that poured in carbon dioxide atmosphere.
Hydrogen	"	Smooth except where stream struck mould.
Coal gas	"	Similar to that poured in hydrogen atmosphere.
"	Blacklead	Smooth. Few splashes at bottom.
"	Lamp-black	Smooth.
"	Lard oil and graphite	Free from oxide skin. Some splashes.

With a low casting temperature or slow rate of pouring, ingots cast without a volatile mould coating show in the upper portion a tendency to a smooth waviness of the surface at the edges.

From the experiments described it may be generally stated that the smoothness of the surface of a chill cast brass ingot depends largely on two factors, viz. (1) absence of surface oxidation, and (2) an initial lag in the transference of heat



(a)



(b)

FIG. 64.

Radiographs of surface layers of cast 70 30 brass ingots, $\frac{1}{2}$ actual size.

(a) Cast with volatile-mould coating, showing subsurface porosity.

(b) Cast with lamp-black mould coating and reducing atmosphere, showing subsurface soundness.

MOULD COATING

from the metal to the mould surface, such as is a thin refractory coating. A third factor is over-heating of the mould surface.

Reference has previously been made to the types of occurring in representative commercial ingots and rolled strip (see p. 14). Fractures and cut sections from ingots show cavities of varying size, roughly spherical in shape, distributed on the cross-section but mainly within about 0.15 inch from the face of the ingot. These are not connected with contraction and are clearly due to the entrapping of gases. As will be shown later, however, gases are not evolved from 70/30 brass on solidification (see Chapter XI, p. 105). The source of the entrapped gases is consequently confined to those injected by the stream of metal during pouring and the gaseous product evolved by the heated mould-dressing. The close proximity of the majority of the spherical cavities to the ingot face suggests the mould-dressing as being chiefly responsible, and this view is supported by the existence, in a number of cavities examined, of carbonaceous material, occasionally in detachable quantities. Further confirmation is given by the casting of ingots in a vertical $12 \times 6 \times 1$ inch mould having one face coated with oil and graphite dressing, the other face being uncoated or covered by a thin sheet of unprotected steel. Sections from such ingots show the presence of small cavities only near the surface which had been cast in contact with the coated face of the mould. Examination of ingots similar to those described in Table IX, made with various types of mould-dressing material, indicates that subsurface cavities are formed only when an inflammable dressing is used.

The most satisfactory means of comparing the soundness of the whole of the subsurface layer of different ingots is the examination by X-rays of a section or slice 0.125 inch thick machined from the ingot, comprising the outer surface and the subsurface layer. This method has been previously referred to in connection with the examination of small portions of commercial ingots (see Chapter III, p. 19). Fig. 64 (a) shows a radiograph of the surface slice of a typical $12 \times 6 \times 1$ inch ingot cast vertically with a volatile mould coating of lard oil and graphite. The small white spots indicate the cavities existing in the ingot within a depth of 0.125 inch from the

surface. Local surface irregularities are seen as small faint grey areas. The number, size and distribution of the cavities vary in different ingots cast by the same method, but the defects shown are characteristic of ingots cast with all the common varieties of volatile mould-dressing and are usually to be found in ingots cast by the ordinary methods described in Chapter II, p. 8. A radiograph of a surface slice representative of ingots cast vertically with an external gas flame and a mould coating of soot is shown in Fig. 64 (*b*). No subsurface cavities or other defects are apparent.

It is thus clear that the beneficial effect of the volatile mould coating in providing a shield of reducing gases, a protection for the mould, and a degree of heat insulation, is accompanied by the disadvantage that the gaseous products do not escape completely through the molten metal but form a source of subsurface defects in the ingot.

In the ordinary method of brass casting from a crucible, the mould is leaned at an angle of about 15 to 30° from the vertical whilst being filled (see Fig. 5). With electric furnaces and mechanised methods of handling the mould, a pouring bowl is essential to enable the charge to be cast direct from the tilted melting furnace, and this has necessitated the filling of the moulds in the vertical position. In the experimental casting of 12×6×1 inch ingots for the separate examination of the various factors involved in the casting method, vertical moulds were adopted in order to avoid any disturbing effects due to contact of the stream with the mould face. The results of the study of mould coatings, however, indicate that the position of the mould may itself be a factor of no small importance as regards the occurrence and distribution of ingot cavities.

While in a vertical mould the gases from the mould coating may escape fairly freely through the metal and between the metal and the mould, in the inclined mould the gases must rise in part through the ingot thickness. This point becomes of particular importance after solidification has commenced on the two faces of the ingot. In the vertical mould the gas bubbles would tend to escape through the still liquid interior, but in the inclined mould, any particles of gas rising towards the "upper" face would be in danger of becoming entrapped against the face of solidifying metal (see Fig. 65).

In general it is found that the lid or ' surface of a brass ingot produces more spill defects in strip than the "back" surface, and this difference is most probably a direct consequence of the inclination of the mould during casting.

These views are confirmed by the features of experimental ($12 \times 6 \times 1$ inch) ingots cast with the mould leaning at an angle of about 30° to the vertical and using an inflammable

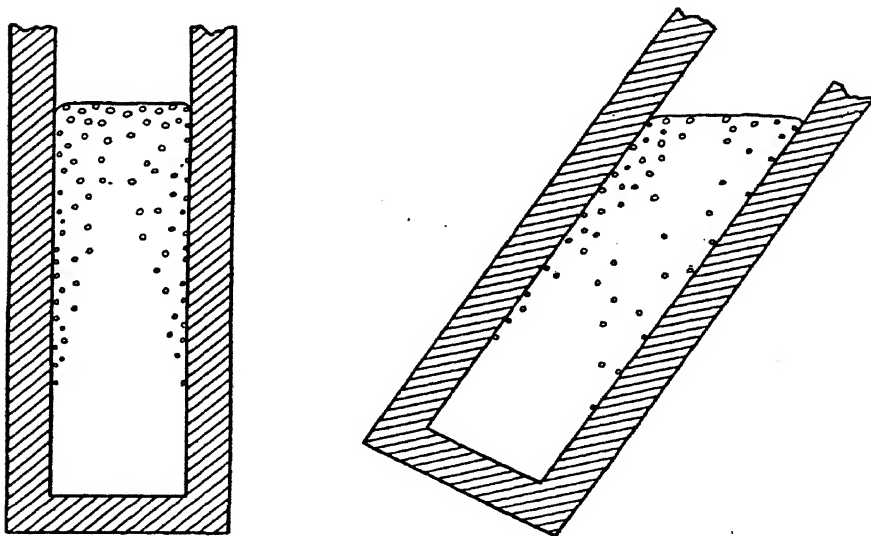


FIG. 65.

Effect of position of mould on liability to entrapping of injected gases.

mould-dressing. Radiographs of surface slices show that large numbers of subsurface cavities are formed near the lid or "upper" surface and only a small number near the back surface. It may thus be regarded as established that the inclination of the mould during casting serves to intensify the bad effects of the inflammable type of mould-dressing.

Quite apart from the effects of the mould-dressing, the method of inclining the mould is considered to be capable of directly and separately introducing ingot defects. An appreciable quantity of the air (or the gases existing above the metal) is introduced into the molten metal within the mould by the

injector action of the stream, and a proportion of this may be retained as inclusions near the upper surface of the ingot similarly to the gases projected into the ingot directly from the coating on the mould wall.

In confirmation, experimental ingots made in an inclined mould without any volatile mould-dressing show, on X-ray examination, that subsurface cavities do exist in the top half of the ingot near the upper face and none near the opposite face.

It is thus evident that the method of inclining the mould is in itself conducive to defects and may intensify the influence of bad casting conditions. The ill effects of the use of too low a casting temperature, for example, would be definitely increased by inclination of the mould during casting.

The manner in which the structure of the ingot and the distribution of cavities are influenced by turbulence in the mould due to the force of the stream has been previously discussed (see Chapter VI). From the nature of the turbulence arising from the use of a volatile mould-dressing, it would appear evident that this is an important additional factor which must be taken into consideration in the interpretation of the structures of commercial ingots. A direct comparison of the structures of two ingots poured under comparable conditions, one with an inert mould coating and the other with an oil "dressing," show that the volatile coating causes a general decrease in crystal size and a definite decrease in the amount of columnar crystallisation. A feature especially noticeable is the increased depth of the outer layer of very fine chill (equi-axial) crystals when a volatile mould coating is used.

Confirmation of the effects described above is given by an ingot in which both of the structures are combined. This ingot, shown in section in Fig. 66, was produced by casting in a mould, only the upper half of which had been coated with volatile material (resin). From a comparison of the two structures it is clear that the effect of the gases evolved from the volatile mould-dressing is to maintain turbulence in the liquid alloy over practically the whole period of cooling to the liquidus temperature, and probably beyond that stage over a considerable proportion of the ingot. At the bottom of a normal ingot the gases from the dressing are evidently completely



FIG. 66.

Section of 70-30 brass ingot cast in mould coated on lower half with lamp-black and on upper half with resin, showing effect of volatile coating on type of crystal formed.

evolved at a comparatively early stage, and the penetration of hot metal from the stream to this region is probably adequate to set up a considerable temperature gradient from centre to surface of the ingot, inducing the growth of columnar crystals. Elsewhere it would appear that the constant motion due to the rise of gases continues sufficiently long to produce throughout the section a nearly uniform temperature, either near the liquidus (giving conditions under which alloys such as 70/30 brass, having a range of solidification, can form equiaxial crystals), or more probably below that temperature, in which case numerous nuclei would already be present. These may have been formed normally in the mass, or originally have become solid in contact with the mould face and later washed into the liquid. The occurrence of a wide "spongy" region would be expected where solidification has been nearly simultaneous throughout a considerable volume of alloy, and for this reason one of the indirect effects of the use of a volatile mould coating is the production of distributed contraction cavities.

In the absence of the effects of a volatile material between the mould and ingot, as in the lower half of the section illustrated, the structure is of the type more readily interpreted on general lines, showing the thin layer of fine equiaxial "chill" crystals adjacent to the mould, the growth of columnar crystals from this layer through a relatively steep temperature gradient, and the central region of equiaxial crystals produced by the combined factors of range of solidification, inoculation with solid particles, and temperature effects of the stream turbulence, as described on p. 58.

In a similar ingot of brass of approximately 60/40 composition, the structure, typical of alloys having only a small range of temperature between liquidus and solidus, is almost wholly columnar over the lower three-quarters of the length, and equiaxial in the remaining top quarter. The contraction cavities are confined to a very narrow central zone. Since in a cooling ingot of an alloy of this type there is only one isothermal surface on which solidification is proceeding, columnar crystallisation can continue however small the temperature gradient, and the tendency to this form of structure is consequently strong. The effect of the turbulence due to gases from the

volatile mould-dressing is, however, evident in the top portion of the ingot, indicating that in this region the gases were not completely evolved before commencement of solidification. This suggests further that with a lower casting temperature and consequent decrease of the time of cooling to the freezing point, the prolonged turbulence would tend still more to inhibit columnar crystallisation.

The relatively long time required for the completion of the gas evolution from the mould-dressing in the casting of brass may be further illustrated by a comparison of the two ingot sections shown in Fig. 67. These ingots ($12 \times 6 \times 1$ inch) were cast from the bottom upwards. Ingot (*a*) was cast in a mould having a volatile dressing, and (*b*) with a mould coating of inert material. The difference in structure between the two bottom-cast ingots is as marked as that between those top-cast, showing that during the filling of the mould the rate of volatilisation of the coating is slow compared with the rise of the metal in the mould.

With increase in the cross-section of the ingot, the time required for the mass to fall in temperature to the freezing point is lengthened, and the production of gases by the mould coating ceases at a correspondingly earlier stage. In the case of 70/30 brass cast at 1100° C. an ingot 2 inches thick shows the effect to a lesser degree than occurs in ingots 1 inch thick, and with increase in thickness to 6 inches the structure is of the normal type, showing no effects of the operation of the additional factor due to the mould coating (see Fig. 57). The indirect effect of the volatile mould "dressing" in producing unsoundness due to contraction cavities is thus confined to thin moulds in which the period of solidification is less than that required for complete volatilisation of the mould coating.



(a)



(b)

FIG. 67.

Sections of bottom cast 70-30 brass ingots showing effect of
mould coating on structure.

- (a) Mould coated with volatile dressing.
- (b) Mould coated with inert material.

CHAPTER XI

ANALYSIS OF THE CASTING PROCESS (*continued*)

The Effect of Dissolved Gases

Porosity Due to Dissolved Gases—Effect of Rate of Cooling—Gas Porosity and Shrinkage Porosity—Variation in Porosity of 5 per cent. Tin—Bronze Castings by Gas Treatments—Absence of Such Effects in 70/30 Brass—Effect of Vapour Pressure of Molten Metal.

Two types of gas cavities in ingots have been already described, and have been attributed respectively to gases from the mould-dressing entrapped near the surface and to gases injected by the stream during pouring. Both of these types of cavity are due to the conditions of pouring, and have been assumed to contain whatever gas constitutes the mould atmosphere in any particular case.

As is well known to all metal founders, however, another type of gas-containing cavity is frequently encountered in some metals and alloys, due to the evolution, during solidification, of gases dissolved in the molten metal. Most metals are capable of dissolving considerable quantities of gases, the solubility being, as a general rule, much greater in the liquid than in the solid state. Under ideal conditions of solidification the gases evolved would escape through the liquid metal, but under normal conditions particles become entrapped. In chill castings the unsoundness due to dissolved gases is not so common as in sand castings, and while this may be partially due to the more ready escape of the gas under conditions favouring columnar crystallisation, it is more probable that, owing to the rapid solidification, the gas is retained in solid solution in the metal.

The gas most commonly responsible for unsoundness is hydrogen, which is soluble to a greater or less extent in all molten metals. In the case of aluminium and its alloys, which are very liable to show "pinholing" due to dissolved

gases, it has been shown¹ that hydrogen enters the metal largely as a product of reaction of the metal with steam; a corresponding degree of surface oxidation occurs due to the accompanying liberation of oxygen in the reaction. The resulting unsoundness in aluminium castings is probably due to the simple evolution of hydrogen as such on solidification, but in certain other metals it is due to a reaction between soluble gases resulting in the evolution of a gaseous product on solidification. In the case of copper, for example, it is known that porosity is caused by a reaction between hydrogen and oxygen (both of which are soluble in the liquid metal), to form steam.² Similarly, sulphur and oxygen together in molten copper cause unsoundness due to evolution on solidification of sulphur dioxide—an effect very noticeable in “blister copper.”

In the case of brasses containing 60 per cent. to 70 per cent. copper the authors have considered that the gas cavities observed are attributable to causes other than the liberation of dissolved gases. Although there is some published evidence to suggest that annealing brasses *in vacuo* permits gas to be drawn off, the results obtained by different investigators vary greatly, owing in many cases to poor experimental conditions, and are of little value.

The authors have investigated the possibility of inducing gas porosity in 70/30 brass castings under conditions designed to give the maximum effect.³ This work was carried out on small cylindrical castings, 6 inches long and 1.5 inches diameter, since it is essential that any comparison of the effect of gas content of different melts of metal on the soundness of resulting castings should be made under controlled cooling conditions. Strip ingots, on account of their rapid and non-uniform rate of solidification, are not a suitable form of casting for such experiments. The influence of variation in rate of cooling can be determined in small castings by varying the mould material, or heating the mould. A satisfactory range is provided by moulds of copper (walls 1 inch thick), cast iron (walls 0.25 inch thick), steel preheated to about 800° C.* and dry sand. Ingots of 70/30 brass poured in 6 × 1.5 inch moulds of these materials at a casting temperature of 1100° C. become

* The steel mould should be protected against oxidation in some way, such as by calorising.

completely solid in times of about 14, 22, 36 and 155 seconds respectively. Density determinations on bars poured from one charge of metal into a mould of each material give the total porosity due to the combined effects of shrinkage and gas evolution for each rate of cooling. If the porosity due to shrinkage can be determined from a comparable series of castings made from gas-free metal, the separate effect of gas absorption can be estimated. The accuracy of this method is ample to provide data for practical use.

Various methods have been proposed for obtaining liquid metal free from gas. Melting and casting *in vacuo* is perhaps the only one which is ultimately completely satisfactory. Presolidification or treatment of the melt with an inert gas is, however, effective in removing a large proportion of the gases present in many molten metals, and in the case of copper, nitrogen treatment is the most satisfactory.^{4, 5} Tin-bronzes are alloys known to be particularly liable to porosity due to dissolved gases, and a comparison of castings of 5 per cent. tin-bronze, made respectively after treatment in the molten state with (a) hydrogen and (b) nitrogen, gives useful indications as to the utility of the method proposed for demonstrating the effect of gases on soundness.

A 5 per cent. tin-bronze melted in a gas furnace, and saturated with hydrogen by passing a rapid stream of this gas through the molten metal for 30 minutes before casting, gives ingots which are notably unsound after any of the four rates of solidification referred to above; the percentage volume of cavities calculated from density determination increases from about 2.5 per cent. when poured in a copper mould to about 10 per cent. in dry sand. Duplicate castings made from metal melted in an electric furnace and through which nitrogen is passed for a uniform period immediately prior to casting, contain about 1.25 to 4 per cent. of cavities (copper and sand moulds respectively). In other words, the amount of internal unsoundness in small cast bars of 5 per cent. tin-bronze, largely if not entirely free from dissolved gas, is increased by 100 to 150 per cent. when the molten metal is allowed to absorb hydrogen.

Similar experiments under identical conditions with 70/30 brass, however, give very different results. The freezing range

of 70/30 brass is about 40°C . as compared with 120°C . for 5 per cent. tin-bronze, and it is therefore to be expected that the bronze castings would give more unsoundness due to contraction cavities. 70/30 brass treated with nitrogen in an electric furnace and cast at 1100°C . in the same 1.5 inch diameter moulds, gives bars containing a percentage volume of cavities of less than 0.5 per cent. in the copper mould, and about 3 per cent. in the sand mould. Treatment of the brass in a gas melting furnace with a stream of hydrogen does not affect the unsoundness of the castings to any appreciable extent, the density values being, within the limits of experimental variation, the same as for brass melted in nitrogen. The amount of unsoundness present is of the order that, from analogy with other similar alloys, may be attributed to shrinkage cavities only. Similarly, treatment with sulphur dioxide, a gas which not only reacts with copper but according to some workers ⁶ is responsible for blow-holes in cast brass, produces no major effect on soundness, although large amounts of sulphide and oxide (presumably of zinc) form in the molten metal and float to the surface as a scum. This inability to vary the unsoundness of cast 70/30 brass by treatment with nitrogen, hydrogen or sulphur dioxide appears to hold for the whole of the range of melting and casting temperatures used in practice, though casting at a low temperature after treatment with sulphur dioxide may be found to give a small increase in unsoundness in rapidly solidified bars due to the entrapping of non-metallic inclusions. In contradistinction to many metals and alloys, brass treated with hydrogen in the crucible gives the cleanest and soundest bars, presumably owing to the absence of oxides.

It is suggested that the factor responsible for this somewhat exceptional behaviour of 70/30 brass is the high vapour pressure of the molten alloy. The partial pressure of zinc vapour in brass varies with composition and temperature as shown in Table X.⁷ Analogy may be made with aqueous gas solutions, in which the solubility of a gas (provided it does not react chemically with water) decreases with increasing vapour pressure of the water, becoming zero at the boiling point. It would appear probable that the solubility of chemically inert gases in brass decreases in a similar manner as the partial

TABLE X
VAPOUR PRESSURE OF ZINC IN THE INDUSTRIAL BRASSES ⁷

Composition.	Partial Pressure of Zinc Vapour at Various Temperatures.				
	Melting Point (Liquidus).		Approximate Casting Temperature.		Boiling Point ° C. (Vapour Pressure Zn 760 mm.).
	Temperature ° C.	Vapour Pressure mm. Hg.	Temperature ° C.	Vapour Pressure.	
Zinc	419.5	0.139	500	1.27	918
60/40	900	160	1040	600	1070
65/35	930	170	1070	595	1100
70/30	955	150	1100	540	1145
80/20	1010	85	1150	265	1300
90/10 (Estimated values)	1055	20	1200	80	1600

pressure of the zinc approaches atmospheric pressure, becoming zero at the boiling point.

It seems reasonable, therefore, to conclude that, whatever the conditions of casting, 70/30 brass is not liable to unsoundness caused by the evolution of dissolved gases. The atmosphere above the molten brass in the crucible consists of evolved zinc vapour, and the partial pressure of any soluble gases is accordingly lowered, so that they are automatically expelled from the metal. The view previously expressed that gas cavities in brass ingots are attributable to the introduction of gases by mechanical means is therefore confirmed. This conclusion is supported by the fact that the use of low casting temperatures in ordinary practice is followed by an increase in the number of spherical cavities. The results of other work further show that the occurrence of such cavities is not a characteristic of the metal itself but can be controlled or entirely eliminated by modifying the casting conditions. While it is considered very improbable that industrial brasses of high zinc content possess any appreciable solubility for gases, it has not been overlooked that some degree of solubility is possible if the remote assumption is made that the solubility in the solid metal is sufficiently high to enable the gases to be retained. The possibility of these conditions occurring appears extremely small. One of the most definite indications of the evolution of dissolved gases during the solidification of an ingot is the rising of the top surface, a common phenomenon in the manufacture of "un-killed" steel and over-poled copper. The fact that this is not observed in the commercial casting of brass ingots is further evidence that effects due to dissolved gases are negligible.

With brasses of high copper content, the partial pressure of zinc at the casting temperature is considerably lower than with 70/30 brasses. 90/10 brass, for instance, which is commonly cast at about 1200° C., has at this temperature a vapour pressure of less than 100 mm. Hg., and the occasional occurrence of central gas cavities in this metal, giving rise to central blisters in rolled sheet, suggests that in such alloys the evolution of zinc vapour may not be always sufficiently rapid to preclude entirely the solution of other gases.

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CHAPTER XII

MATERIALS FOR THE CONSTRUCTION OF INGOT MOULDS

Present Practice and Recent Developments—Properties of Grey Cast-iron Moulds—Causes of “Blowing”—Warping of Steel Moulds—Temperature Gradients in Mould Wall of Cast Iron, Steel and Copper—Properties Desirable in a Mould Material—Effect of Mould Material on Ingot Quality.

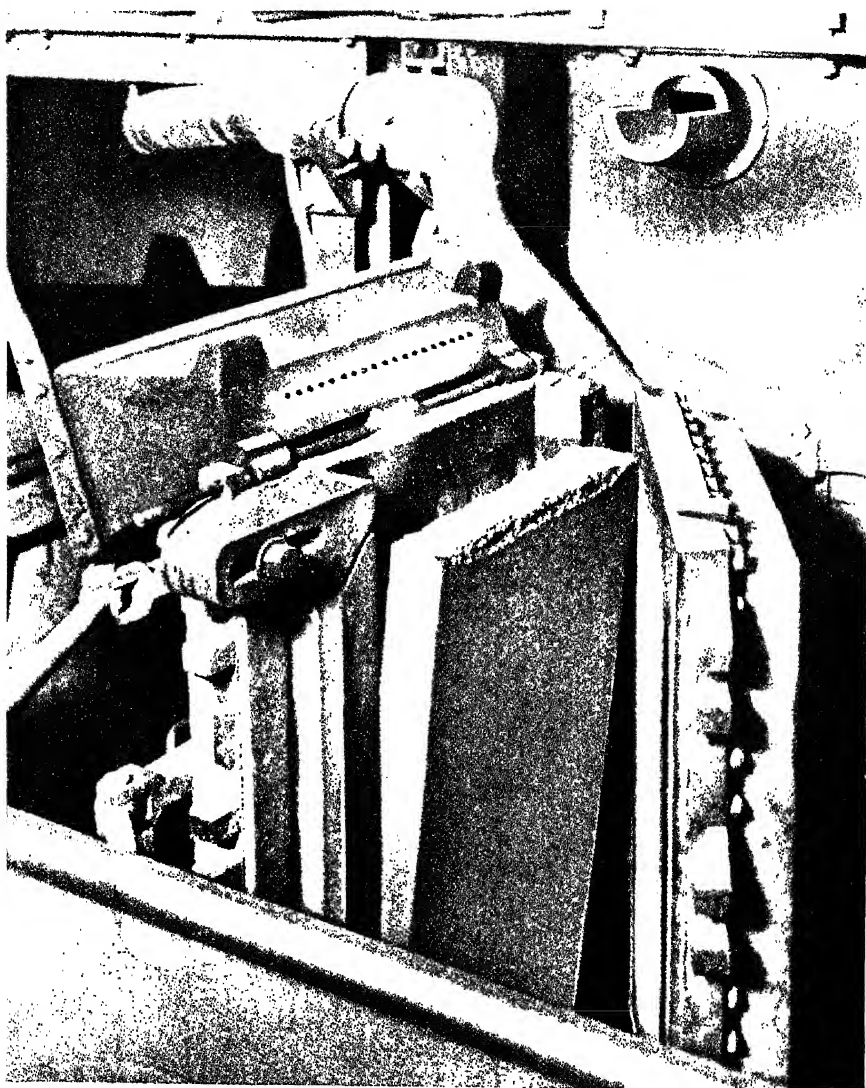
GREY cast iron was introduced to replace the earlier stone moulds more than a century ago, and since that time has been in almost universal use as a mould material for the casting of all metals. Cast iron offers the great advantages of durability, cheapness and ease of casting to shape with a smooth surface. It cannot be considered an ideal mould material, however, since it is subject to the formation of cracks on the inner working faces, frequently at an early stage in its life. A further troublesome characteristic of cast iron is the tendency to evolve gases when overheated, producing the ingot defect known as “blowing” (see p. 16). An abnormal example of this defect in an ingot of 70/30 brass is shown in Fig. 68. It occurs in particular at the point of impingement of the stream on the mould face during casting, and is not generally serious in brass manufacture by commercial methods which include the use of a mould-dressing of volatile carbonaceous material, but with higher melting-point alloys such as “nickel-silvers” containing 25 to 30 per cent. nickel, the trouble is more frequently encountered.¹ The thermal insulating effect of a lamp-black (soot) mould coating is less than that of the volatile type and is often insufficient to prevent local overheating and blowing defects in brass, unless precautions are taken to avoid impingement of the stream.

Little scientific work has been done on the fundamental properties of ingot moulds, or on specific points affecting the use of cast iron for this purpose. In fact it can safely be said

PLATE XXXVIII.



FIG.
: poured in cast-iron mould at initial temperature of 300 C.
showing extensive "blowing" defects.



[Courtesy of Messrs. I. C. I. Metals, Ltd.

FIG. 69.

Junker copper-faced water-cooled mould opened for removal of ingot.

that cast-iron moulds have been accepted over many years without the fundamental properties requisite in a mould material having been adequately examined. Certain individual properties of cast iron of interest in this connection have, however, been investigated. Pearce and Morgan ² have given considerable information on the thermal properties of ingot mould irons, and Donaldson (*loc. cit.*) has investigated thermal conductivities, while Brearley ³ discusses in some detail the design and treatment of cast-iron ingot moulds used for steel casting. Much of this discussion is not applicable to non-ferrous work, but, on the other hand, such points as cleanliness and freedom from cracks of the mould surface, the advisability of a slight camber on the ingot, etc., are equally important in all types of ingot casting. Matuschka ⁴ has studied the thermal changes during the cooling of a steel ingot in the mould and has determined the temperature at different positions in the wall of a mould of commercial size.

The subject of ingot moulds for non-ferrous work has received more attention as a result of the introduction by Junker of copper-faced water-cooled moulds for the casting of brass strip ingots. These moulds consist essentially of thin copper plates (0.25 to 0.75 inch in thickness) which form the working faces of the mould and are bolted on to a steel or cast-iron shell which permits cooling of the backs of the plates by circulating water. The essence of the method, according to British Patent Specification No. 237,622, is the use of water cooling of thin metal plates, which are insulated from the main part of the mould structure, the insulating material also serving as an effective packing to maintain a tight joint between the copper plate and cast-iron casing. The mould stands vertically and the two faces are hinged at one side to open in book fashion, thus permitting easy handling (see Fig. 69); the circulating water is admitted through the lower hinge and allowed to escape through the upper hinge. Internal baffles are arranged to ensure an adequate circulation of the water over the surface of the copper plates. The inventor claims ⁵ not only freedom from cracking of the working faces such as is experienced with cast iron, but longer life, more rapid working, and ability to control the cooling rate of the ingot by regulation of the water supply.

Rohn ⁶ has also used with considerable success copper-faced water-cooled moulds of special design, for the casting of nickel alloys of high melting point and claims a long life for these moulds even when surface dressings are entirely dispensed with. Roth ⁷ has discussed in general terms the application of water-cooled moulds to brass casting. Copper-faced water-cooled moulds are stated by Oertel ⁸ and by Hessenbruch and Bottenberg ⁹ to be successfully used for the casting of steel billets.

In more recent years an entirely new type of thin plate water-cooled mould has been advocated by Erichsen. In this mould the plates are of an alloy of the "Invar" type having a low heat conductivity (British Patents Nos. 358,697, October 1931, and 299,850, May 1929). The features of these recent developments and their value from the industrial and metallurgical points of view are discussed in Chapter XIV (p. 141).

The range of composition of sixteen samples of cast iron used for ingot moulds is shown in Table XI. The greatest variation is in the phosphorus content, which may be high in poor quality irons. Within these limits of composition no marked differences in tendency to blowing are detectable, and this difficulty with cast iron is therefore probably not related to composition.

TABLE XI

VARIATIONS IN COMPOSITION OF CAST-IRON INGOT MOULDS

	Per cent.
Graphitic carbon . . .	2.5 to 3.1
Combined carbon . . .	nil to 0.8
Silicon	1.5 to 4.0
Manganese	0.3 to 1.0
Sulphur	0.04 to 0.13
Phosphorus	0.14 to 1.2

The difficulties of overcoming blowing with cast-iron ingot moulds and of judging on existing data the potential value of new materials suggested for mould construction, have been considerable owing to the absence of knowledge regarding

the causes of blowing and the fundamental properties of the mould.

Blowing is due obviously to a gas evolution at the metal mould interface, since surface cavities are produced in the brass. It is liable to occur with any graphitic cast-iron mould, while materials such as dead-mild steel, copper and pure graphite are free from this trouble under the ordinary conditions of service. It would appear from this that the mould material itself is the responsible factor, and that the gases are not evolved from a metallic reaction between the brass and the mould faces. That the presence of graphite itself or the possible porosity of the surface layers due to the mechanical removal of the graphite are not the major causes of "blowing" is, however, shown by the fact that cast irons of widely different structures varying from grey to white, but of similar carbon content, give the effect to approximately the same extent. High-carbon steels containing up to about 1 per cent. carbon also behave in this respect in a manner similar to cast irons.

For the experimental investigation of the tendency of different mould materials to give blowing, the authors¹⁰ used the small-scale apparatus described on p. 91, whereby a small disc of molten brass was poured on to a horizontal plate of the mould material under examination. The results of such tests show that, with high-carbon steel or cast iron, gas cavities are formed in the underside of the disc (provided no insulating surface coating is applied to the plate) in all cases where the plate surface is oxidised, and that increase in the degree of oxidation increases the volume of such cavities. The results of some tests made in this way are shown in Table XII.

With mild steel containing less than 0.1 per cent. carbon no appreciable gas evolution occurs under ordinary pouring conditions whatever the degree of oxidation. If, however, the ingot is poured in a mild steel mould which is initially at a high temperature (*e.g.* red heat), the time of solidification is so much prolonged that sufficient gas pressure may be generated at the steel-oxide interface to cause some of the thick oxide scale formed on the mould at this temperature to be pressed into the ingot surface. Reduction of the carbon content of the surface layers of white, and to a lesser extent of grey cast irons by

annealing in oxide, decreases but does not eliminate the tendency to blowing.

TABLE XII

THE INFLUENCE OF DIFFERENT DEGREES OF OXIDATION AND CARBON CONTENT ON THE "BLOWING" EFFECT OF IRON-CARBON ALLOYS ON DISCS OF 70/30 BRASS

Plate.	Surface of Plate.	Temp. of Testing, ° C.	Degree of Oxidation.	Volume of Gas Holes in Brass Disc. C.c.
Grey cast iron	Machined	100	Nil	0.17
		320	Blue	0.29
		650	Black	0.65
Mild steel 0.1% C.	Machined	100	Brown	0.01
		400	Deep blue	0.01
		700	Black	nil
High-carbon steel 1% C.	Machined and ground	100	Nil	nil
		100	Pale straw	0.04
		350	Straw	0.10
		350	Deep blue	0.20
White cast iron	As cast	350	Straw	0.09
		350	Deep blue	0.14

It thus appears that "blowing" is definitely due to a surface reaction in the mould between an oxide film and carbon or carbide in the mould material, and the volume of gas evolved (presumably oxides of carbon) increases directly with both the carbon content and the degree of oxidation.

The temperature at which such a reaction occurs in a cast iron and the degree of surface oxidation necessary can be determined by heating samples of cast iron *in vacuo* and measuring the volume of gas evolved at different temperatures. Machined rods of grey cast iron, heated *in vacuo*, are found to give off a considerable volume of gas between 700° and 800° C., the evolution becoming rapid at 750° C. (e.g. in a typical experiment 1.4 c.c. at N.T.P. per sq. cm. of surface was evolved on heating to 800° C. for two hours). If such a treatment is repeated several times the gas evolution ceases, but on withdrawing the sample from the furnace and allowing it to stand in air, further considerable gas evolution occurs at the same

PLATE XL.



FIG. 70.

Structure of typical grey cast iron. $\times 250$.



FIG. 71.

Central area of grey iron with original structure as shown in Fig. 70, after treatments *in vacuo*, showing deposition of graphite. $\times 250$.



FIG. 72.

Edge of cylinder shown in Fig. 71, showing decarburisation of pearlite. $\times 250$.



FIG. 73.

Structure of typical non-pearlitic cast iron. $\times 100$.

temperature on reheating (e.g. 0.7 c.c. at N.T.P. per sq. cm. under the same conditions). Although surface oxidation may not be apparent, the necessity for some oxidation of the iron to occur before gas evolution can again take place is shown by the fact that allowing it to stand in perfectly pure dry nitrogen and then reheating, does not result in any further evolution of gas. White iron specimens treated in the same way give similar results and repeatedly evolve considerable volumes of gas between 700° and 800° C., but only when some oxidation between annealings is permitted.

Micro-examination of grey cast iron so treated shows the reaction to involve the combined carbon of the iron rather than the graphite. In one series of trials using a sample containing originally 2.85 per cent. graphite and 0.55 per cent. of combined carbon, and with the structure shown in Fig. 70, the authors found that after repeated oxidations and annealings *in vacuo* a considerable deposition of graphite from the carbide had occurred throughout the bar, giving a structure in the central areas of graphite, pearlite and ferrite, as shown in Fig. 71. In the surface layers, however, no pearlite was visible, the structure being entirely ferrite plus graphite (see Fig. 72). At the same time analysis showed that the graphite content of the centre of the bar had increased by 0.15 per cent. at the expense of the combined carbon, while in the surface layers similar increase in graphite content had occurred with almost complete removal of the combined carbon.

From the fact that the combined carbon is largely involved in this reaction, it may be thought that a non-pearlitic cast iron with a structure consisting entirely of ferrite and graphite would be free from blowing, but while with such an iron the effect is undoubtedly much reduced it is still liable to occur. This may be due to some reaction with graphite itself or to porosity where the graphite has become removed mechanically from the surface layers. The use of a non-pearlitic iron with a structure such as that shown in Fig. 73 is therefore not entirely satisfactory in overcoming blowing, and in any case is a somewhat soft and mechanically unsuitable material for ingot moulds.

In actual practice the presence of an oxide film on the surface of a cast-iron ingot mould is very difficult to avoid entirely.

The thickness and nature of the usual type of mould-dressing employed in the works provide some protection against oxidation as well as against local overheating during pouring, but such protection cannot be relied upon, and in ordinary commercial use cast-iron ingot moulds cannot be treated in a manner which will ensure the entire absence of blowing under severe conditions. The only method, therefore, of obtaining freedom from this trouble lies in the use of a carbide-free mould material which will not give off gas on reheating even when superficially oxidised. It has already been stated that dead-mild steel fulfils this requirement, but moulds of this material are found in practice to be open to two serious objections. The first is a tendency for the stream of metal to weld on to the mould face at any point where impingement occurs, and the second is a tendency for the mould to warp so seriously as to become useless after a small number of casts. Fig. 74 illustrates the nature and amount of distortion encountered in a 24-inch long strip ingot mould of mild steel (0.1 per cent. carbon) in which only six heats had been poured.

The change in form produced in a mass of material such as an ingot mould by non-uniform heating and cooling is due to thermal stresses arising from variable changes in volume through its section. Other factors being equal, the thermal stresses and resulting distortion are higher the greater the variation in temperature in the mould wall.

Accurate measurements of the temperature gradients in ingot moulds of different materials during the casting and solidification of an ingot of 70/30 brass have been made by simultaneous measurements of the temperature at different positions in the mould wall throughout the operation,¹⁰ under conditions of pouring ensuring that the metal stream was clear of the mould face. The temperature gradients in moulds of mild steel, cast iron and copper respectively at various intervals of time from the completion of pouring a 1-inch thick ingot in a mould with walls also of 1 inch thickness are shown in Figs. 75 and 76. It will be seen from these curves that with mild steel and cast-iron moulds the inner surfaces are heated rapidly by the molten metal to about 600° C. before the body of the mould rises in temperature. After a short interval of

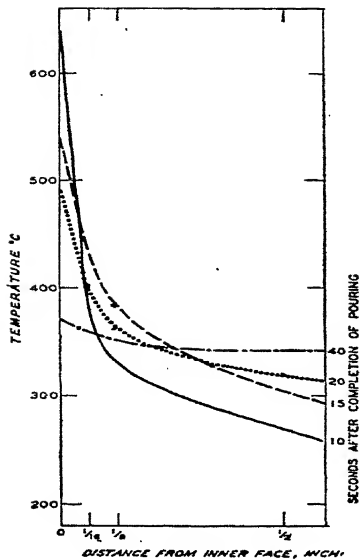


FIG. 75.

Temperature gradients in mould walls of 1 inch thickness on pouring 12 x 6 x 1 inch ingots of 70/30 brass. Mild-steel mould.

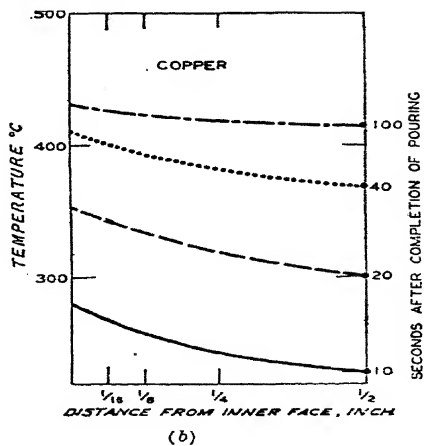
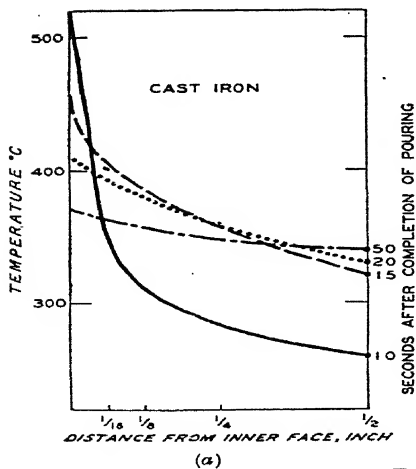


FIG. 76.

Temperature gradients in mould walls of 1 inch thickness on pouring 12 x 6 x 1 inch ingots of 70/30 brass.

- (a) Cast-iron mould.
- (b) Copper mould.

time a stage is reached (corresponding possibly to the shrinking away of the ingot from the mould) when the conduction of heat through the mould wall exceeds the rate at which heat is supplied to the inner face by the ingot. The temperature of the inner surface layers then falls while that of the main body of the mould rises. In the case of copper, conduction of heat through the mould wall is so rapid that there is no great difference in temperature through the wall at any time.

If the stream of metal is allowed to impinge on the surface of a cast-iron mould during pouring, the temperature of the surface layer at this point rises to about 800° C., which is sufficiently high to permit a rapid evolution of gas owing to the reaction previously discussed. A comparison of the temperature gradients in such a cast-iron mould with and without impingement of the stream on the mould face is shown in Fig. 77.

The rapid changes in temperature in the inner surface layer of a steel or cast-iron mould cause large and rapidly changing stresses in this portion of the mould wall. The inner face expands rapidly during the ten seconds immediately after pouring. The mould tends to open at top and bottom to accommodate this increase in length, but is held by the clamps and by the rigidity of the thick mass of cold metal at the back. In the hot layers of the mould a severe compressive stress is thus set up which is largely relieved by permanent deformation. The inner surface layers then cool while the main body of the mould is heating up. The stresses at this point are reversed and a tensile stress is set up on the inner face. When the mould is released the elastic stresses set up by deformation and subsequent cooling cause the mould to warp, and become concave on the inner face as shown in Fig. 74. Cast iron cannot flow plastically to any appreciable extent and the tensile stress on the inner face causes transverse cracks to appear. The presence of these cracks prevents warping, further stresses being relieved by a slight opening of the transverse cracks, and progressive deterioration produces a condition of the inner surface familiar to users of such moulds (see Fig. 78).

In the case of copper the thermal stresses set up within the mould wall are very small, and entire absence of elasticity in

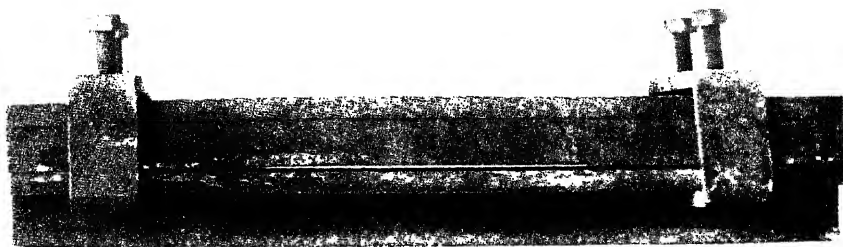


FIG. 74.

Mild steel ingot mould after six heats of 70, 30 brass.

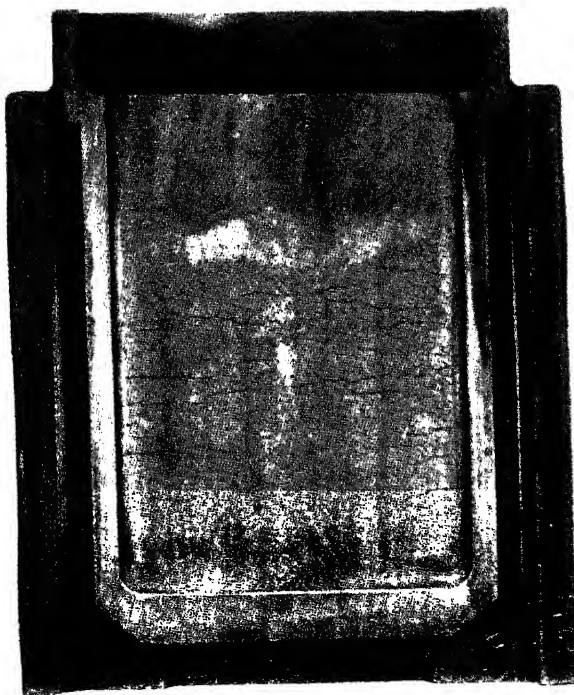


FIG. 78.

Inner face of grey cast-iron ingot mould after approximately 100 heats, showing transverse cracks.

annealed copper, together with the high ductility of the metal, enables them to be relieved completely by plastic flow during the cooling of the mould.

It would appear from the observations described that the following properties are necessary in a mould material to minimise distortion: (a) a high thermal conductivity, to

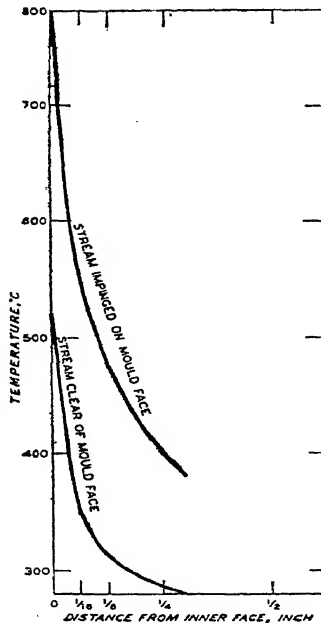


FIG. 77.

Temperature gradients showing the high temperature reached by the mould face on casting due to impingement of the metal stream.

reduce thermal stresses; (b) a high elastic limit, especially at raised temperatures, to minimise plastic deformation, or (c) a negligibly low elastic limit, to permit the entire relief of stresses by plastic deformation.

In ferrous metals the thermal conductivity is always low and a completely satisfactory material for ingot moulds can only be sought in steels having a high yield point, such as heat-treated nickel-chromium-molybdenum steels. Such

materials, apart from the initial expense, are still open to the objection that the stream of metal is liable to weld on to the mould during casting, and there seems no reasonable prospect that such steel moulds would be completely satisfactory in practice. Experimental moulds of heat-treated nickel-chromium-molybdenum steel (yield point 55 tons per square inch) have, however, been made in which the tendency to welding on to the ingot was overcome by aluminising the inner surface, and such moulds proved reasonably satisfactory. With copper moulds, however, the difficulty of welding does not occur provided that the cooling is adequate, and from the economic point of view generally this material possesses the most promising qualities.*

The essential feature in the satisfactory use of copper moulds for the casting of high melting point alloys is the cooling of the inner face. If a 1-inch ingot of brass is poured in a copper mould with walls only 0.5 inch in thickness, and the stream of metal during pouring impinges on the mould wall, the whole thickness of the mould wall at this point will be heated to a high temperature and the ingot will probably "weld on." Cooling of the back of the plate with water overcomes this difficulty, but the same end can to a certain extent be achieved by the use of a copper plate of sufficient thickness to ensure that the material as a whole does not rise to too high a temperature. As an example, a copper mould with a wall thickness of 1.5 inch will have sufficient margin of heat capacity to absorb the heat communicated from a cast ingot of 70/30 brass 1 to 1.5 inches in thickness without rising in temperature to an extent liable to cause welding. The essential factor in the successful use of copper is the high thermal conductivity of the material, and for this reason high conductivity copper should be used.† Whether water-cooling is used to keep the temperature of the whole mould wall sufficiently low or whether an extra thickness

* Aluminium moulds have been used experimentally for the casting of brass ingots, but were not entirely satisfactory owing to the low melting point of the metal. For the casting of metals of low melting point, however, aluminium may prove to be a suitable mould material.

† Impurities have a marked effect in lowering the thermal conductivity of pure copper. For example, at 20° C. the conductivity of copper is lowered about 50 per cent. by the addition of 0.1 per cent. phosphorus, 0.4 per cent. arsenic or 1.5 per cent. tin.

of copper is employed to give a similar effect by heat absorption depends entirely on the conditions of operation, the organisation of the foundry, and the particular requirements concerned.

With steel and similar low conductivity materials the inner face is heated rapidly to a high temperature before the body of the mould wall is heated by conduction, and increasing the thickness of the mould wall above a certain minimum does not, therefore, avoid the high initial temperature of the mould face. With copper, however, the uniformity of temperature under the conditions of the experiments described indicates that conduction of heat through the mould walls keeps pace with the rate of heat input, and increasing the thickness of the mould wall would, therefore, be expected to decrease the maximum temperature attained.

When considering the use of copper for moulds, however, the composition of the alloy being cast, the casting temperature, the method of pouring with particular reference to the tendency for the stream to impinge on the mould wall, etc., must be taken into consideration, and it must be emphasised that solid copper moulds should only be used in lieu of water-cooling where the conditions are favourable for maintaining the inner temperature of the mould wall below red heat during pouring.

For large thick ingots a mould wall of at least similar thickness will probably be necessary and in the casting of alloys of high melting point, such as nickel alloys, water-cooling of a copper mould offers decided advantages, both in manipulation and economy.

A further point to be considered is the influence of the mould material on the properties of the cast ingot. A comparison of the quality of 70/30 brass ingots poured in moulds of grey cast iron, water-cooled copper, solid copper or special alloy steel, does not show large differences in surface, structure and soundness such as might be expected from the variable properties of these materials. As regards surface quality the most important factor is blowing, which is avoided when materials other than cast iron are used. Copper moulds produce a slightly greater superficial chilling effect if the mould-dressing employed

is extremely thin, but the use of even a moderately thick coating of lamp-black is sufficient to neutralise this effect and to give surface of high quality.

Internal soundness as measured by density determinations on complete ingots made in each type of mould shows no appreciable variation and, given the same casting conditions and size of ingot, the effects produced by using a copper mould in place of one of ferrous material are negligible from this point of view. Although the use of a copper mould does produce some change in ingot structure, the effect is not great.

On general grounds it may be imagined that the use of an ingot mould of highly conductive material would increase the temperature gradients in the ingot during solidification and increase the tendency to columnar crystallisation, but this effect is not large in strip ingots of the order of 1 inch in thickness. Figs. 79 and 80 show the structure of $12 \times 6 \times 1$ inch ingots of 70/30 brass cast in grey cast-iron and water-cooled copper ingot moulds respectively, an inert mould-dressing of lamp-black being used to eliminate the disturbing factor of turbulence due to the mould-dressing. The ingot cast in the water-cooled copper mould shows the formation of somewhat longer columnar crystals; otherwise the structures are similar. Larger ingots ($24 \times 6 \times 1$ inch) cast in solid copper and heat-treated alloy-steel moulds show a similar small difference in the same direction. It is clear that other factors such as turbulence in pouring have a much greater effect, though with larger ingots of wider section it is probable that more distinct differences may be obtained by the use of a more conductive mould material. In strip ingots of thin section it does not seem probable that any appreciable variations in structure could be introduced by variations in the rate of flow of the cooling water in a copper-faced water-cooled mould.

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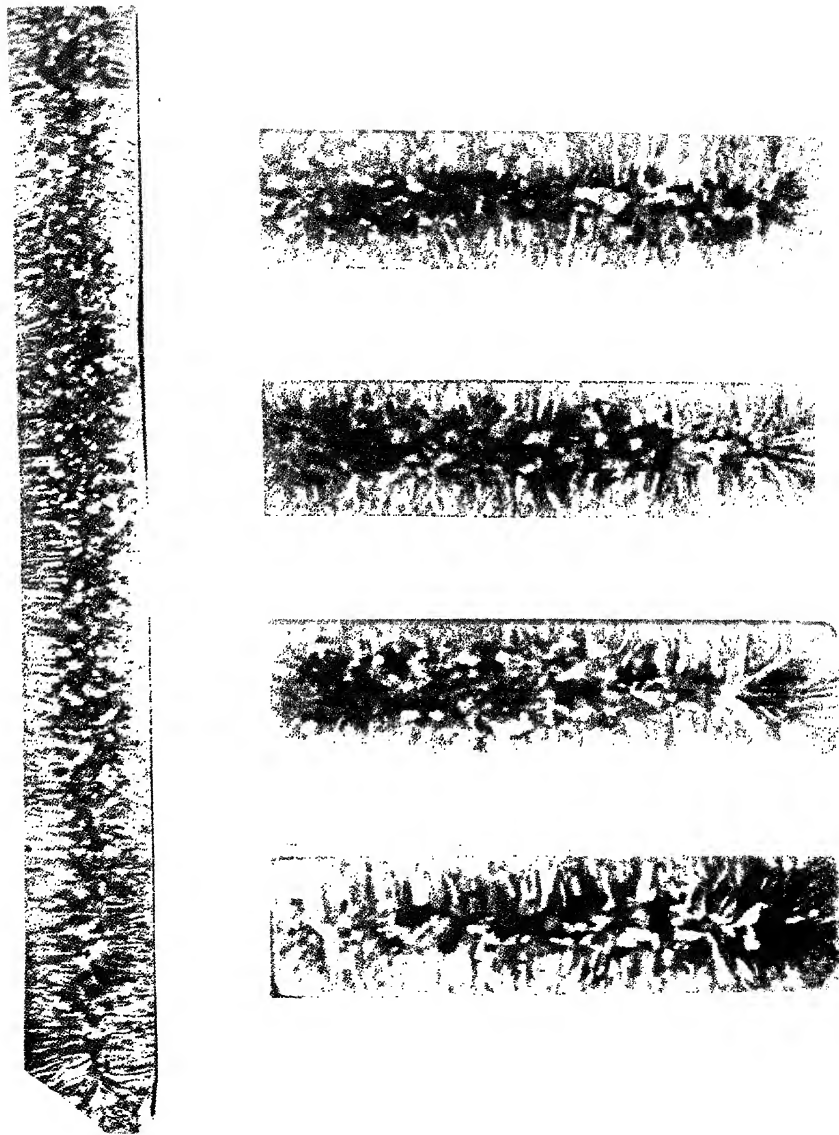


FIG. 79.

Structure of 70, 30 brass ingot cast in grey iron mould coated with lamp-black.

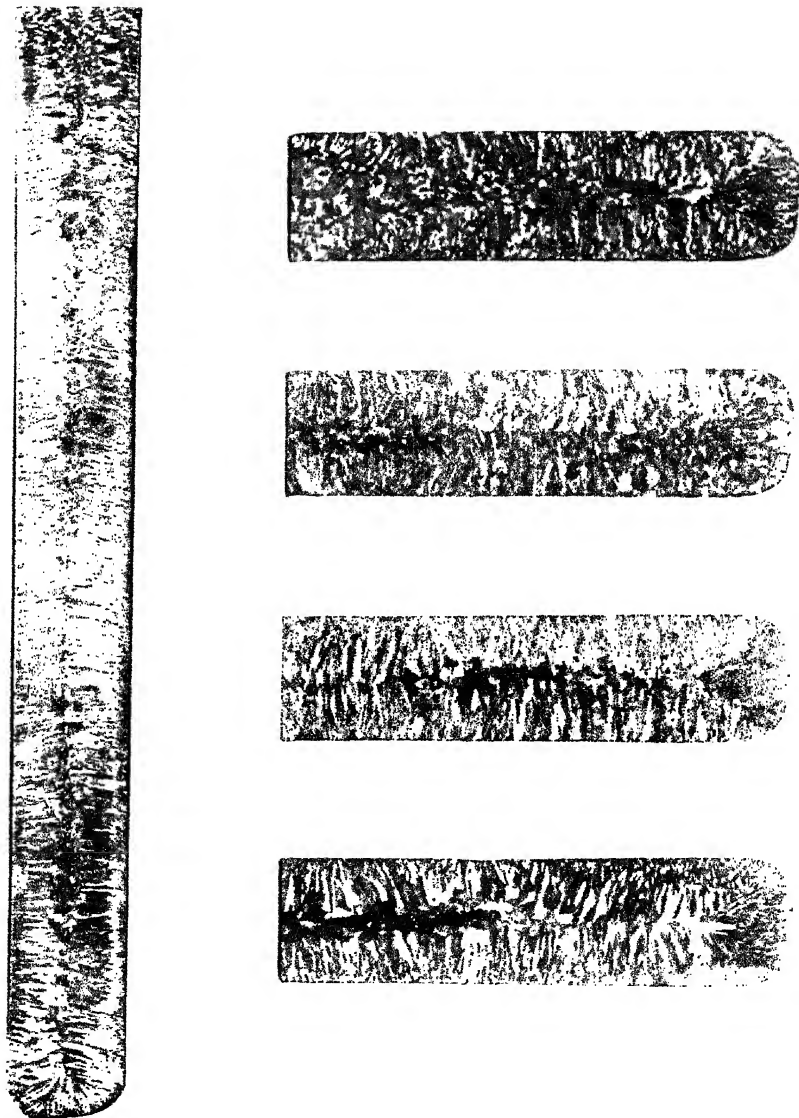


FIG. 80.

Structure of 70/30 brass ingot cast in water-cooled copper mould coated with lamp-black.

- ⁵ O. Junker, *Z. Metallkunde*, 1926, 18, 312; *Metallwirtschaft*, 1930, 9, 406.
- ⁶ W. Rohn, *Z. Metallkunde*, 1927, 19, 473-78.
- ⁷ W. Roth, *Giesserei*, 1933, 20, 413.
- ⁸ W. Oertel, *Stahl und Eisen*, 1929, 49, 696.
- ⁹ W. Hessenbruch and W. Bottenberg, *Mitt. K.W. Inst. Eisenforschung*, 1931, 13, 205.
- ¹⁰ G. L. Bailey, *J. Inst. Metals*, 1932, 49, 203.

CHAPTER XIII

MODIFIED AND IMPROVED CASTING METHODS

Most Favourable Conditions—Recommended Improvements in Ordinary Practice—Alternative Methods Eliminating Volatile Mould Coatings.

WHILE a number of practical recommendations have been made as they have arisen during the discussions presented in the foregoing chapters, it is at this stage convenient to consider the data and information collectively, and to summarise the more outstanding points of practical importance in order to deduce recommendations for the improvement of the ordinary methods of brass casting and the possible introduction of new methods.

As a result of the examination of the individual influences of the numerous factors involved in the casting process, the formation of the ingot during casting and solidification can be visualised and the temperature gradients and oxidation effects at any stage can be roughly predicted from the known conditions present in any particular case. Since the production of an ideal temperature gradient, that is from bottom to top of the ingot, is impracticable with slab ingots of ordinary form, the only method of securing freedom from contraction cavities would consist in the use of an extremely low casting temperature. This is not practicable in view of the inevitable introduction, under these conditions, of surface defects and cavities due to entrapped gases. In general, therefore, the casting conditions favouring central soundness (freedom from contraction cavities) are disadvantageous from the point of view of improvement in the quality of the strip ingot as a whole. This type of unsoundness, however, is rarely large in amount, and being always situated near the middle of the thickness, appears to have relatively little effect on the quality of rolled strip. The most promising line of improvement of

the material consequently appears to be the adoption of the conditions most favourable to the elimination of subsurface cavities and surface defects such as folds in the ingot.

The conditions essential for this purpose in ordinary commercial casting methods are :

(a) A high casting temperature and a rapid rate of pouring, the main effects of which are to minimise the risk of entrapped gases and to avoid surface folds. The central contraction cavities which are inevitably produced by such conditions are reduced in extent by the use of measures to lessen the degree of penetration of the stream within the mould, and for this purpose the adoption of a pouring bowl or tundish is beneficial.

(b) The use of the minimum amount of volatile mould coating necessary to give freedom from surface oxidation effects. The connection between the volatile mould coating and the presence of subsurface cavities in the ingot indicates the desirability of decreasing as far as possible the volume of gas produced on the mould face. Further, the thinner the mould coating the more rapidly are the gases liberated.

(c) Casting with the mould placed vertically. The method of casting in an inclined mould is conducive to the entrapping of the gases from the mould atmosphere by injection, as well as of gases evolved directly from the mould coating. Provided that the mould is vertical it may be filled either from a multiple-hole tundish or by a single stream.

Any of these modifications singly or combined will have an influence in the direction of increasing the freedom of the ingot from defects which affect the quality of rolled strip.

Other refinements which have an important effect in improving the quality of the ingot by avoiding defects due to the mould and the inclusion of slag with the stream consist in the use of copper moulds, either solid or water-cooled, in place of cast iron in order to obviate possible blowing defects, and the employment of a bottom-pouring ladle which ensures a clean stream of metal with simple means of regulating the pouring speed. The extent to which such additional measures can be adopted will vary with different works.

The modifications described above represent improvements which can be introduced into the ordinary method of casting without altering the main features of the method or the plant

employed. In view of the advances which will probably be made in the hot rolling of brasses of 70/30 type and other alloys at present cold rolled, reference may be made to the advantages of using ingots of shorter length and greater thickness. Apart from the metallurgical advantages of thick ingots, the moulds are more stable when used for vertical casting, and the number of moulds to be stored, handled and "dressed" is economically reduced. The method of inducing directional solidification by thermally insulating one edge of the mould and pouring near to the insulating slab (described in Chapter IX, p. 80) has not been applied in moulds of larger dimensions than $12 \times 6 \times 1$ inch. By employing the principle of this method, it is probable that the central soundness of ingots could be definitely improved, but it is doubtful whether, in industrial practice, the advantage would be commensurate with the economic and manipulative difficulties involved.

Whilst the relatively simple measures outlined above are capable of improving ingot quality and reducing wastage to an important degree, it cannot be hoped that the ingot ideally free from subsurface cavities will become a regular product unless steps are taken to avoid all means of gases becoming entrapped in the ingot. In view of the inevitable injector effect of the stream it would appear that in the chill casting of long thin ingots the entrapping of minute particles of air is not completely avoidable by any means where, in pouring, a stream of metal penetrates the liquid ingot in the mould. The most prolific source of entrapped gas cavities in commercial brass ingots, however, is the commonly used volatile mould coating, the effects of which are intensified when the mould is filled in an inclined position. Since the known functions of the mould coating do not necessitate the production of gases between the metal and the mould walls, but consist mainly in the avoidance of oxidation and the protection of the mould surface, it should clearly be possible to devise various means by which the volatile type of mould coating can be eliminated and its functions fulfilled by alternative means. The experiments with cast discs (described on p. 95) and the further experiments on small ingots cast in reducing atmospheres provide a basis for a method of this kind.

By fixing a gas burner below a bottom-pouring ladle in such

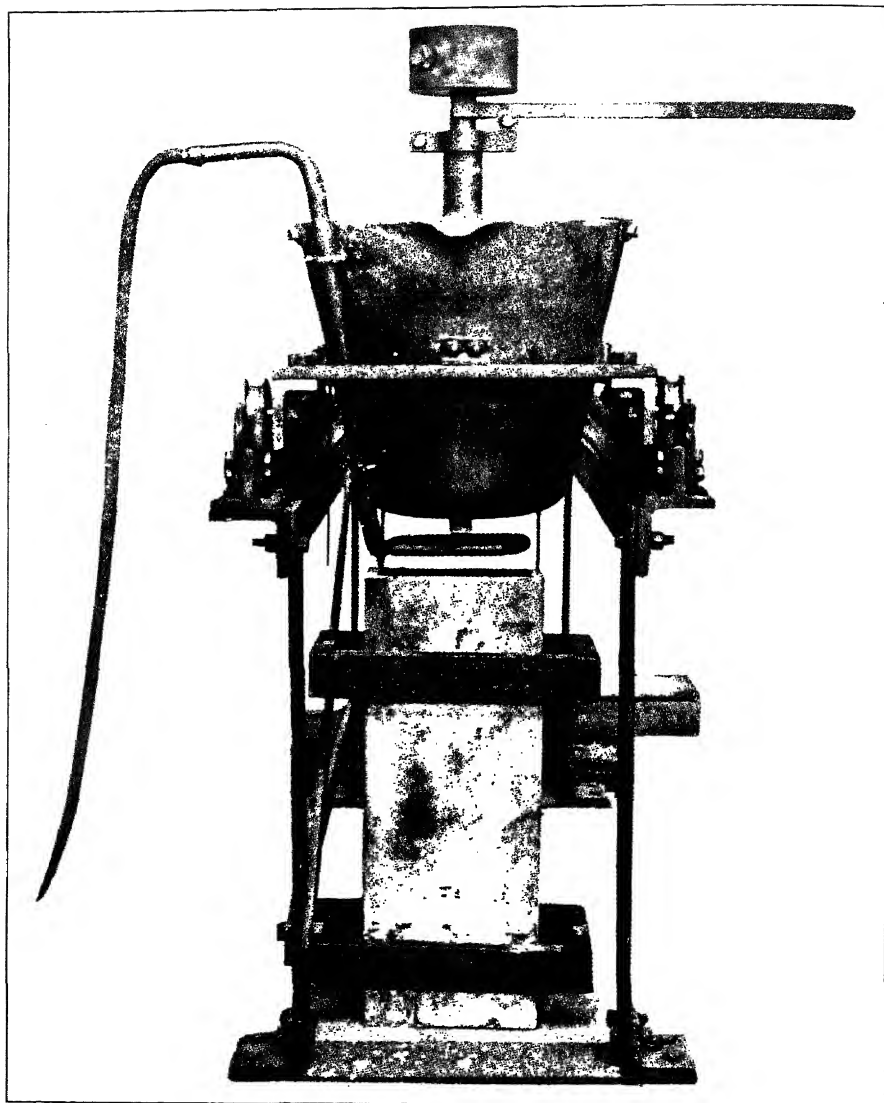


FIG. 81.

Experimental apparatus for casting brass ingots, 24 6 1.5 inches by improved method, using solid copper mould, inert mould coating, reducing atmosphere of gas and bottom-pouring ladle.

a way that the entire space between the mould and ladle is filled by a luminous flame of coal gas, oxidation of the stream of metal can be entirely avoided; during pouring the stream is visible through the flame, appearing orange in colour and free from the white flame of burning zinc vapour characteristic of brass poured in air. An apparatus devised for the experimental casting of strip ingots $24 \times 6 \times 1.5$ inches by this method is shown in Fig. 81.

The ladle consists of an iron body (as normally supplied for foundry ladles) lined with a refractory material. A large crucible cut down to a suitable length forms a good inner lining which can easily be cleaned. In the base of the ladle is fixed a moulded carborundum nozzle made by ramming carborundum cement in a steel mould or die. The stopper of the ladle by which pouring speed is controlled consists of a heat-resisting steel tube to the lower end of which is fixed a point of "salamander" shaped to fit the nozzle. The gas ring, which can be adjusted in position below the nozzle of the ladle, is constructed from a suitably bent pipe pierced with numerous holes along the inside of the ring. The metal is transferred from the crucible to the ladle after rapid skimming (which need not be complete, since slag and oxides rise to the top of the metal when in the ladle and do not enter the stream). The mould is fixed so that the stream can fall clear of the faces. A hot dozzle or feeding-head of fire-clay is placed in the top of the mould before casting and with long and wide moulds a tundish or pouring bowl is also used. This apparatus gives all the conditions for casting at high temperature and high speed into a vertical mould with freedom from oxidation, without the use of a volatile mould coating. Some coating is necessary to protect the mould faces and to give a sufficient degree of heat insulation to avoid too severe an initial chill effect on the turbulent metal surface. Carbon (soot) deposited by an acetylene flame played on the mould surface is a satisfactory material. Tests of the method made by casting 70/30 brass in moulds of cast iron, malleable cast iron, steel and copper, and rolling the ingots to strip, show the material to be of high quality provided the ingot surface is not impaired by blowing. To obviate this danger, which is increased in the absence of a volatile coating, copper moulds are recommended. A number of ingots so produced and

examined by X-rays showed no subsurface unsoundness. These trials have been confirmed from ingots rolled in works and by the comparison of large quantities of such ingots with those of current commercial quality. A casting temperature of 1100°C . and a pouring speed of about 1.5 inches of ingot length per second are suitable for ingots having dimensions of the order of 8 inches in width, 1.5 inches in thickness and up to 24 inches in length.

The reducing gas flame and its regulation constitute one of the chief details of this method and an adequately copious flame is essential to successful casting. The effectiveness of the flame can readily be judged during casting by the appearance of the stream. Occurrence of any oxidation is indicated by the white flame of burning zinc. The crude form of gas ring shown in the illustration, although satisfactory for use with high-pressure gas (pressure 12 inches of water), is not entirely suitable where only low-pressure gas (3 inches of water) is available, and other types of reducing flame may be preferable.

A number of analyses made on samples of gas from flames given by different forms of burner are given in Table XIII. Items 1, 2, 3 and 5 are the mean figures from twelve different estimations all closely agreeing. The figures do not represent the true analysis of the flame atmospheres, since no allowance is made for water vapour or free carbon, but some indication of the proportions, particularly of water vapour, may be gathered from the values of methane and hydrogen compared with those in the original gas. The analyses, however, show relatively the oxidising or reducing characters of the flame atmospheres.

The results show the paraffin burner to be unsatisfactory on account of the high oxygen and low hydrocarbon and hydrogen contents.

The low-pressure burner, an illustration of which is shown in Fig. 82, gives between the mould and ladle an atmosphere nearly approaching the composition of the original coal gas, and is for this reason even more satisfactory than the high-pressure-gas ring burner. The low-pressure burner is constructed so that the coal gas is deflected down into the mould, the top of the mould being protected from air currents by an inverted trough placed over the burner and provided with a

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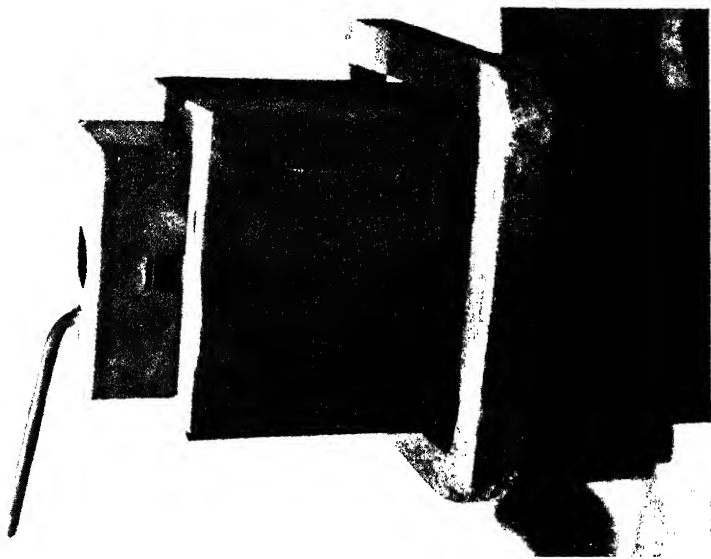
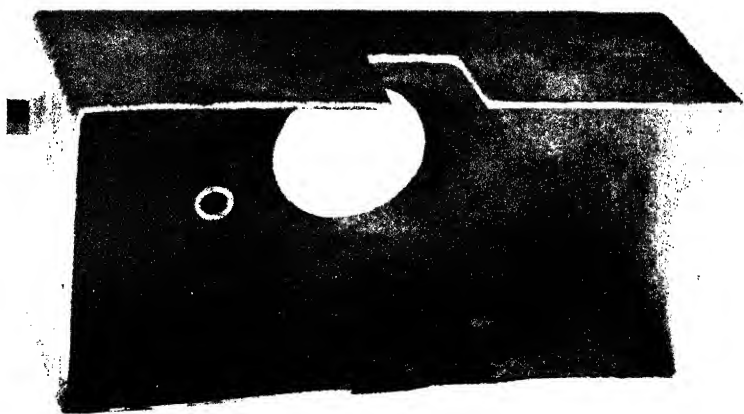


TABLE XIII

ANALYSIS OF FLAME ATMOSPHERES BETWEEN BOTTOM POURING
LADLE AND MOUTH OF MOULD, USING DIFFERENT TYPES
OF BURNER

Item.	Burner.	Analysis of gas : volume per cent.					
		CO ₂ .	O ₂ .	CO.	Hydro- carbons. CH ₄ .	H ₂ .	N ₂ (by diff.).
1	Composition of coal gas used.	4	0.7	9	27	45	14
2	Ring burner (see Fig. 81). High-pres- sure gas.	4	1	6.9	14	25	49
3	Luminous flame from open tube (10 inches from tube mouth). High-pressure gas.	4.6	0.8	7.4	7.8	17.5	62
4	Luminous flame from open tube (19 inches from tube mouth). High-pressure gas.	6	1.5	4.9	1.4	5.5	80.7
5	Burner for low-pres- sure gas (see Fig. 82).	3.6	0.9	8.4	19	36	32
6	Wells paraffin burner (5 lb. pressure).	8.5	4	5.5	1	2.5	78.5
7	Wells paraffin burner (12 lb. pressure).	5	4	10.5	2	7	71.5

hole for pouring. It is, of course, important that the supply pipes and taps are adequate to give the full gas pressure at the outlet pipe (0.375 inch internal diameter in this case). This design has given the most satisfactory results in practice; it is simple to use and ingots can be obtained entirely free from oxide scale whether low- or high-pressure gas is used. Analyses of gases from the bottom of the mould 24 inches in length on samples withdrawn 1½ minutes after the flame had been started (Table XIV) show definitely the superiority of the low-pressure burner over the high-pressure ring burner.

In view of the above data, the influence of time of burning upon the mould atmosphere is of interest. Samples drawn through a short length of fine silica tubing inserted in a hole drilled within one inch of the bottom of a 2-ft. mould at varying

TABLE XIV

ANALYSES OF ATMOSPHERE AT BOTTOM OF 2-FT. MOULD,
1½ MINUTES AFTER LIGHTING COAL GAS BURNER

Burner.	CO ₂ %.	O ₂ %.	CO%.
Ring burner (high-pressure gas) .	4.2	4.3	3.6
Low-pressure burner . . .	4.6	0.8	8.0

intervals after lighting the low-pressure gas burner gave on analysis the results shown in Table XV.

TABLE XV

ANALYSES OF ATMOSPHERE AT BOTTOM OF 2-FT. MOULD.
BURNER AS IN FIG. 82. GAS PRESSURE 3 INCHES
OF WATER

Time of burning before sampling.	CO ₂ %.	O ₂ %.	CO%.
1 second	2.5	12.0	2.2
4 seconds	4.4	4.5	4.0
30 "	5.0	1.4	7.8
2 minutes	4.6	1.0	7.8
4 "	3.6	0.8	8.9

These show clearly that, with the low-pressure burner arrangement, oxygen is rapidly eliminated from the bottom of the mould, the major portion being removed within 30 seconds, using gas at a pressure of 3 inches of water. It is probable that even an ill-fitting mould would give satisfactory results with a burner of this type.

The modified method described involves a considerable departure from ordinary practice, including the elimination of the volatile mould coating and its substitution by an inert coating of lamp-black together with the provision of a reducing coal-gas atmosphere. This method is capable of giving ingots of the highest quality obtainable by top pouring, though, as has been mentioned, the entrapping of gases injected by the stream is still liable to occur, and ingots poured by this method are, as with all other methods of top pouring, liable to give small blisters on cross rolling to sheet.

Reference has previously been made to the fact that the addition of a small percentage of phosphorus (about 0.05 per cent.) has the effect of avoiding the formation of any coherent film of oxide on the surface of molten brass. Although the rate of vaporisation of zinc from phosphorus brass is actually greater than from pure brass, the oxide formed is incapable of being held in suspension or attached to the surface of the molten metal. Brass containing phosphorus can thus be poured with full exposure to air without the use of a volatile mould coating or the alternative gas flame, with no risk of the production of defects due to oxidation. The principle of the process is generally analogous to that of the gas flame method and similar results are obtained as regards quality of the ingot produced. A further characteristic of brass containing phosphorus is that, owing to the apparent enhanced fluidity of the metal, the ingot invariably shows a very clear impression of the mould. Associated with this property, however, is a greater tendency to splashing. This, although without effect in ingots cast in moulds of cast iron or other material of low conductivity, may affect the surface of ingots cast in copper moulds where any splashed metal is rapidly chilled on contact with the mould face. The use of phosphorus addition to the brass together with an inert mould coating such as soot has the great merit of simplicity, and might with advantage be adopted to some extent as an alternative to the gas flame method or that generally used industrially. The principle of the method is that complete means of avoiding defects due to oxidation are present in the metal itself.

The presence of even a small amount of phosphorus affects the mechanical properties of brass, the most noticeable effect being a reduction of grain growth on annealing. This question is discussed in Appendix D, "The Phosphorus Brasses."

CHAPTER XIV

SPECIAL CASTING METHODS

The Durville Process—Erical Process—Bottom Casting.

CONSIDERATION has so far been given primarily to the ordinary crucible casting process for the production of brass strip ingots and to modifications of this process which do not involve any radical departure from the ordinary methods of pouring. Apart from such improved methods, special casting processes have recently been introduced which involve principles novel in the brass foundry. Chief among these are the Durville casting process, which introduces an entirely new method of pouring; the Erical process, for which novel effects are claimed as a result of the use of a special type of mould; and bottom casting, which, although used to a considerable extent in steel practice, has not as yet been applied industrially to the casting of brass and other similar materials.

The Durville Process.

The Durville process * was devised in order to avoid the surface defects to which alloys containing aluminium (such as aluminium-bronze) are liable when poured in the usual way. The tendency of the alumina film to form folds and splashes during pouring has been referred to previously (p. 37), and the essential feature of the Durville process is the avoidance of turbulence during pouring, which is the main cause of these defects. The apparatus used consists of an ingot mould connected by means of a trough to a ladle in such a manner that the lower edge of the ladle, the base of the trough and the lower edge of the mould are in a straight line. The ladle is filled with metal which is then transferred smoothly to the mould by inverting the apparatus. The level of the metal at successive stages in the process of inversion is illustrated

* P. H. G. Durville (British Patent 23,719, 1913).

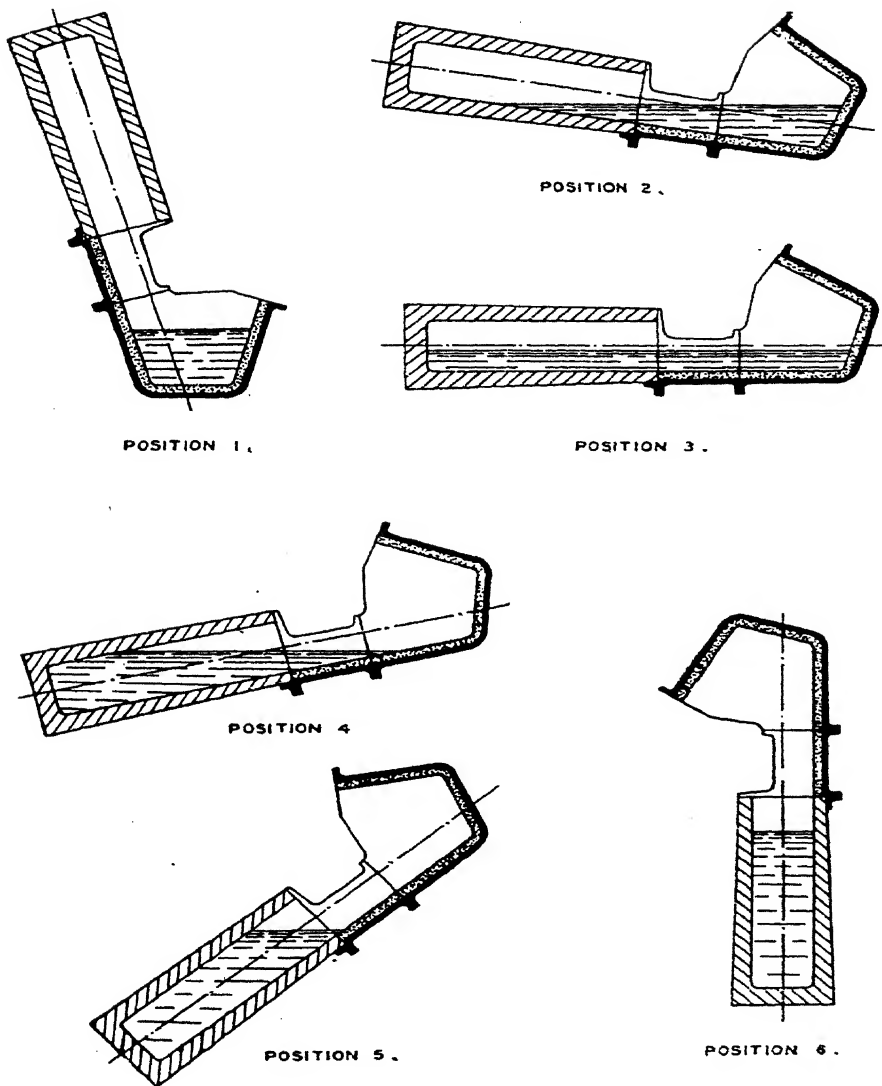


FIG. 83.

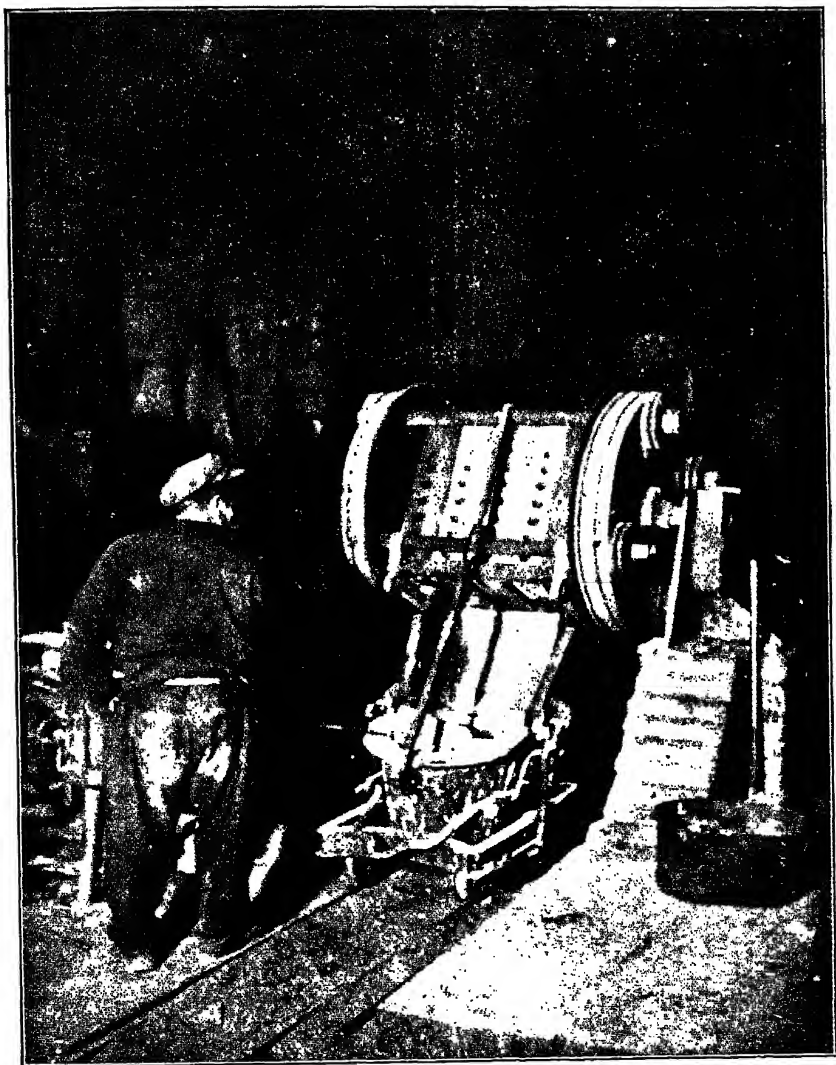
Position of metal surface in Durville casting process at various stages of rotation.

diagrammatically in Fig. 83. The process is in some respects similar to that commonly used in the casting of aluminium strip ingots in which the mould is tilted at an angle, the metal poured down one edge and the mould gradually moved into the vertical position as the pouring proceeds.

The Durville process was originally used for the casting of aluminium-bronze containing 8 to 9 per cent. aluminium for French coinage. The original apparatus designed by the inventor was arranged to cast simultaneously four ingots each approximately $24 \times 14 \times 1.5$ inches in a mould built up with cast-iron plates separated by U-shaped pieces which formed the bottoms and edges of the moulds. The surfaces forming the inner faces of the mould were smooth machined with no further preparation, and no mould-dressing was used. Attached to the top of this quadruple mould was an iron frame which held four separate feeding heads or dozzles of about 2 inches in depth and 1.5 inches in width. These dozzles were built up of asbestos board and refractory brick wrapped in thin sheet iron. Attached to the dozzle frame was a trough about 3 feet in length which connected the ladle to the mould, the inner surface of this trough being of sheet iron covering a layer of refractory brick. The ladle was basin-shaped, having sloping sides lined with refractory and carrying fittings for bolting the mould, trough and ladle together.

In casting aluminium-bronze the metal is melted without the use of any flux and the temperature raised to 1200° to 1300° C. It is then poured into the ladle, which is attached to the casting apparatus as shown in Fig. 84. The metal is skimmed carefully and allowed to cool to about 1100° to 1150° C., the apparatus being then rotated. This operation is smooth and regular and occupies about 30 seconds (see Fig. 85). After detachment of the ladle, the four ingots can be lifted out of the mould, the walls of which have a very slight taper outwards towards the top (see Fig. 86).

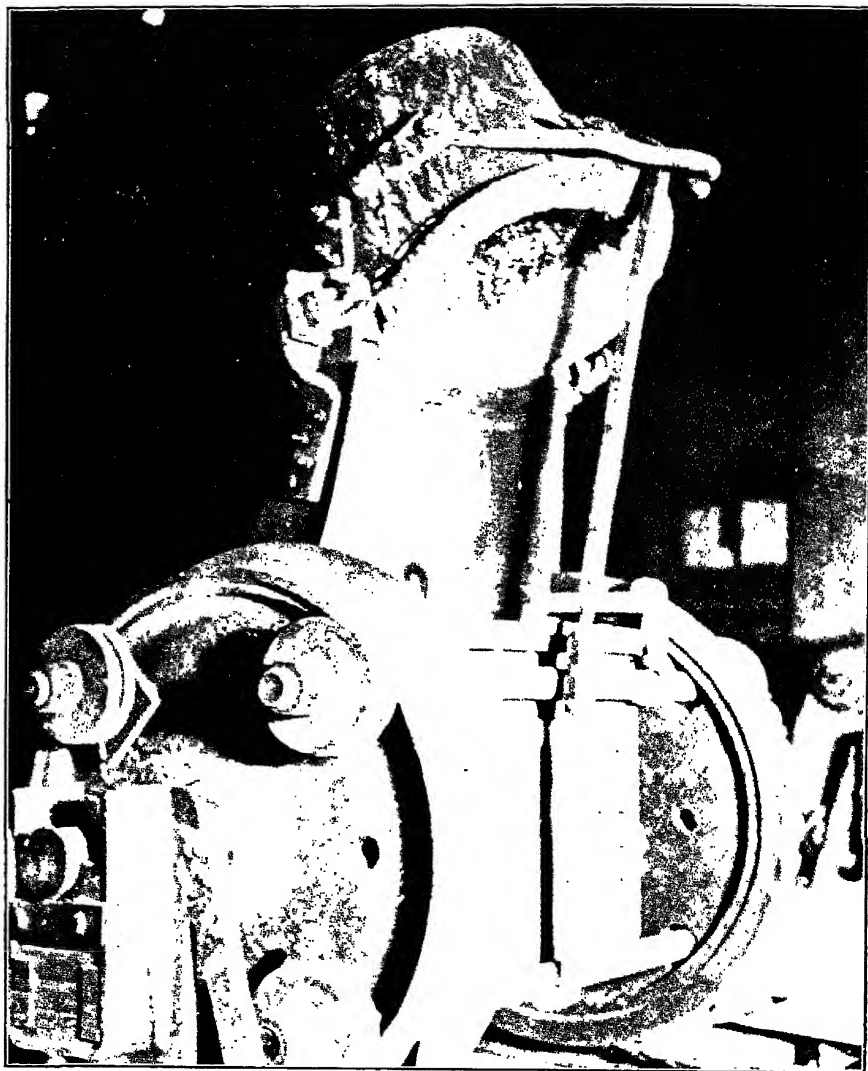
Ingots of aluminium-bronze and of brasses containing aluminium, cast in this way are of high quality. The surfaces, of which a typical example is illustrated in Fig. 87, are smooth, showing only small ridges corresponding to the successive levels of the metal in the mould during rotation. On account of the low casting temperature used, contraction cavities are reduced



[Courtesy of Société des Alliages et Bronzes Forgables.]

FIG. 84.

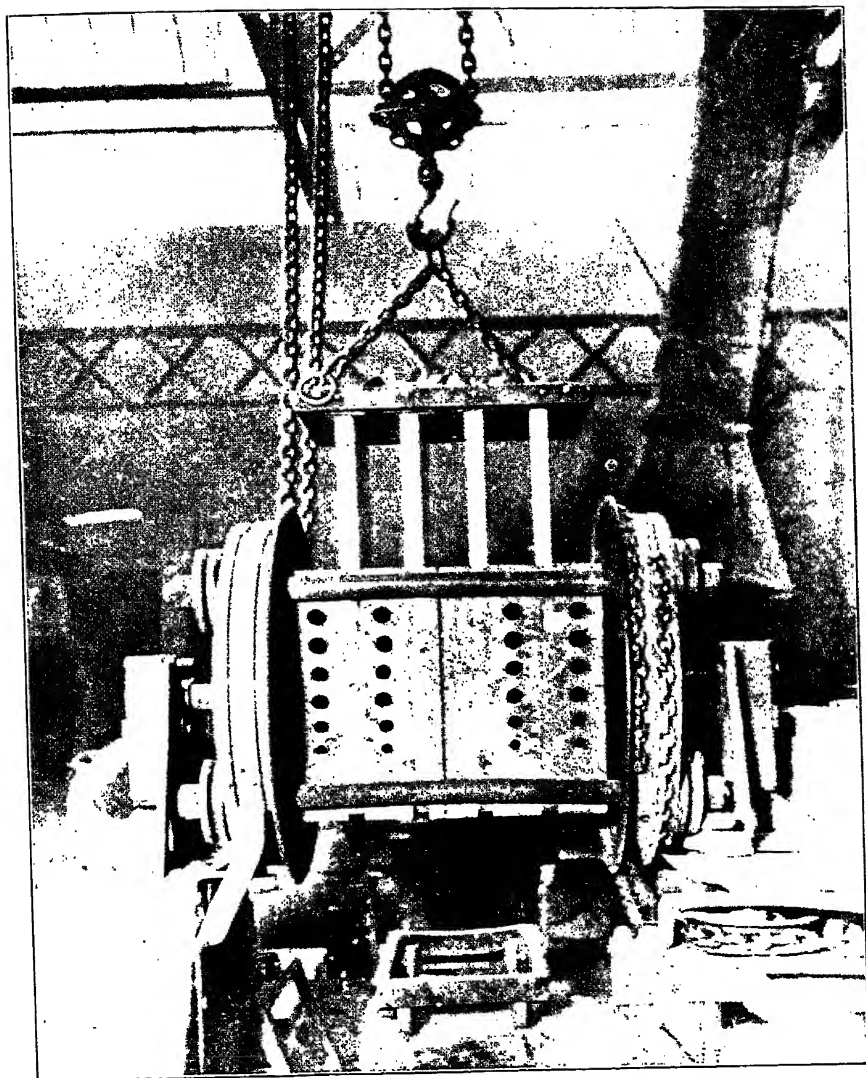
Original Durville machine, ladle attached ready for rotation.



[Courtesy of Société des Alliages et Bronz

FIG. 85.

Original Durville machine. Rotation completed.



[Courtesy of Société des Alliages et Bronzes.]

FIG. 86.

Original Durville machine. Withdrawal of ingots.

PLATE XLIX.

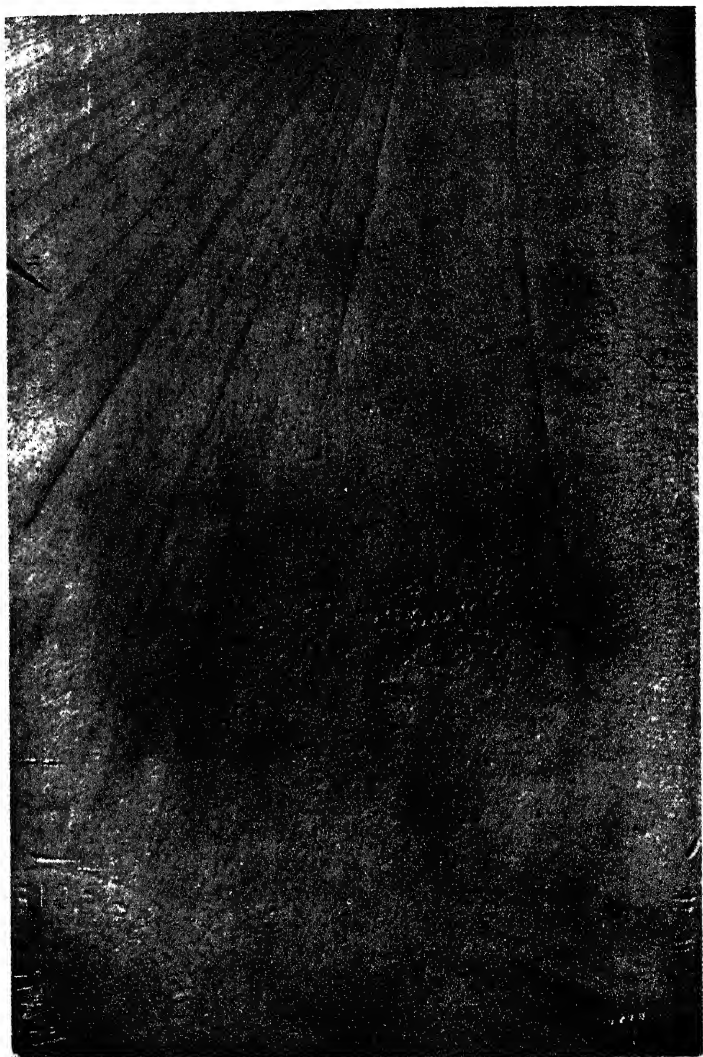


FIG. 87.

Surface of 91/9 aluminium bronze ingot cast by Darville process.

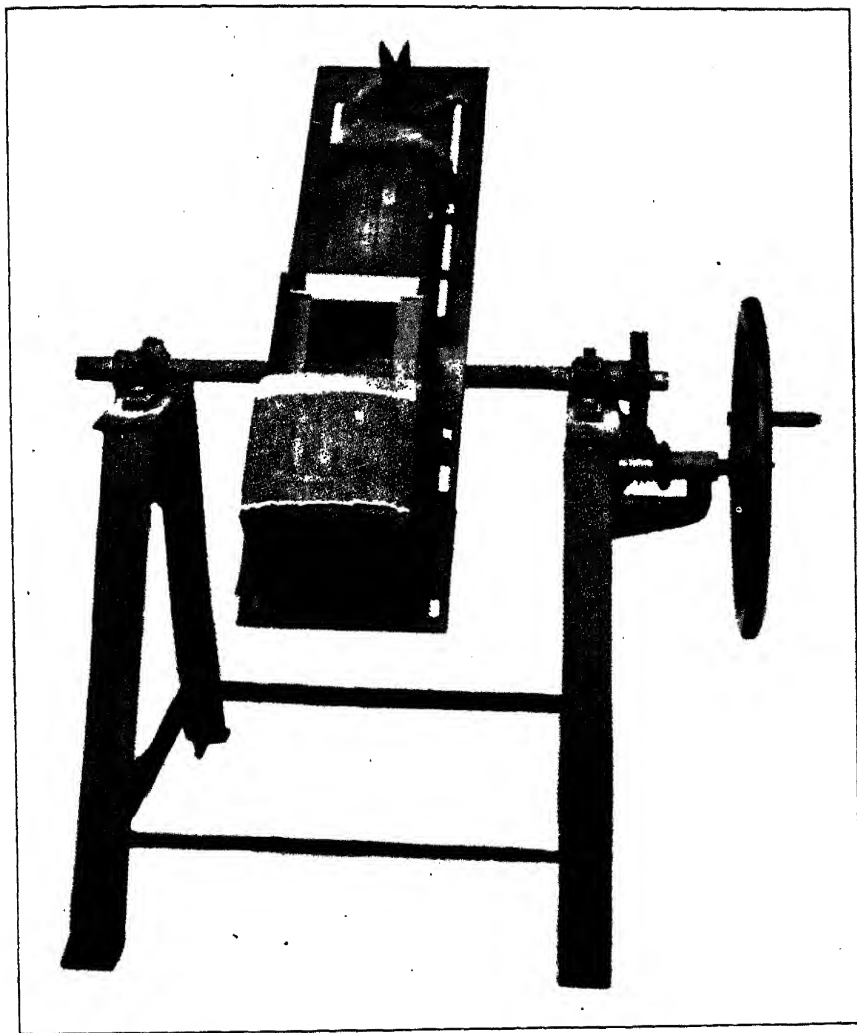


FIG. 88.
Experimental apparatus for Durville casting

PLATE LI.

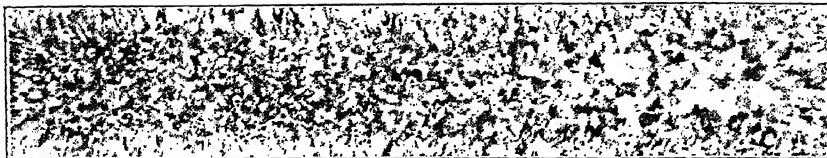


FIG. 90.

Structure of 70/30 brass (0.2 per cent. aluminium) cast by Durville process at 1100°C.
Actual size.

to a minimum, and in general it can be said that such ingots are of high quality as regards surface and internal soundness.

In view of the promising features of this process the determination of the most satisfactory casting conditions for any particular alloy, and the extension of its use to materials other than aluminium-bronze or brass are of considerable interest. For the fuller investigation of the possibilities of the method the authors used the experimental small-scale apparatus illustrated in Fig. 88. This consists of a small ladle and launder cut in one piece from a salamander crucible, with a mould built up of steel plates to give three ingots each $9 \times 4.5 \times 1$ inch. A hot dozzle is used between the mouth of the ladle and the mould.

If pure 70/30 brass is cast in this way with the metal surface freely exposed to air, the skin of zinc oxide which forms is found to break and fold during the casting operations and give rise to surface defects. The presence of aluminium in the brass overcomes this difficulty by producing a strong surface film of alumina; as little as 0.1 per cent. aluminium in 70/30 brass is sufficient to enable the material to be cast with high surface quality, without the use of any mould coating.*

As would be expected from the absence of turbulence during pouring, the tendency to the formation of columnar crystals on solidification is much greater than with ordinary methods of crucible pouring, and can only be inhibited by careful control of the process. At a high casting temperature $9 \times 4.5 \times 1$ inch 70/30 brass ingots containing 0.2 per cent. aluminium show a structure almost entirely columnar whether the speed of rotation is rapid (6 seconds) or slow (24 seconds). As the casting temperature is lowered an increase in the proportion of equiaxial crystals is obtained in rapidly poured ingots, and with a casting temperature of 980°C . (as compared with 1100°C . used for crucible casting) and rapid rotation (6 seconds), the largely equiaxial structure shown in Fig. 89 is obtained. Even with this low casting temperature, however, a slow speed of rotation (24 seconds) will produce a structure entirely columnar, similar to that obtained with high casting temperatures (see Fig. 90). The microstructure of typical portions of surface

* The effect of additions of aluminium on the physical and mechanical properties of the copper-zinc alloys is dealt with in Appendix C, "The Aluminium Brasses," p. 164.

layers of the ingots illustrated in Figs. 89 and 90 are shown in Figs. 91 and 92. The straight crystal boundaries in the surface of an ingot of columnar structure are clearly shown, and the danger of such a structure from the point of view of rolling, has been referred to earlier (p. 64). The effect of varying casting conditions on the ingot surface is small, and only at very slow speeds of rotation are surface laps of any serious magnitude formed.

In the pouring of 70/30 brass containing a small amount of aluminium, the most satisfactory casting conditions from the point of view of both surface and structure are therefore a low casting temperature and a rapid speed of rotation. It is interesting to note that there is no difficulty in casting aluminium brass at a temperature of only 40° C. above its freezing point (liquidus). The ingots are almost invariably free from the numerous cavities of entrapped gas which would inevitably occur in top-cast ingots poured at such a low temperature (see p. 72). Casting at a low temperature has other advantages, chief among which are the reduction of piping, some economy in fuel consumption and the comparatively long period of time during which the metal can remain in the ladle for skimming.

While these observations refer particularly to $9 \times 4.5 \times 1$ inch ingots, parallel conditions have been found to be the most satisfactory for the casting of brasses containing aluminium in larger strip ingots up to 100 lb. in weight, and of cylindrical billets of considerable size for tube extrusion.

The same conditions are suitable in the casting of aluminium-copper alloys by the rotatory process. Aluminium-bronze strip ingots which are entirely columnar in structure are liable to crack in rolling, particularly if the material contains included alumina or other impurity, as is possible when remelted scrap is used. Accordingly, the casting temperature must be rigidly controlled and must be as low as is possible without leaving an excessive amount of metal behind in the ladle. Pyrometric control is advisable. Experimental ingots of 91/9 aluminium-bronze which have been made entirely from remelted scrap have been found to be of good quality for rolling, provided the casting temperature did not exceed 1100° C.* A typical

* For cold rolling it is important that the aluminium content should not exceed 9 per cent.

PLATE LII.



FIG. 91.

Microstructure of surface layer of ingot shown in Fig. 89. $\times 100$.

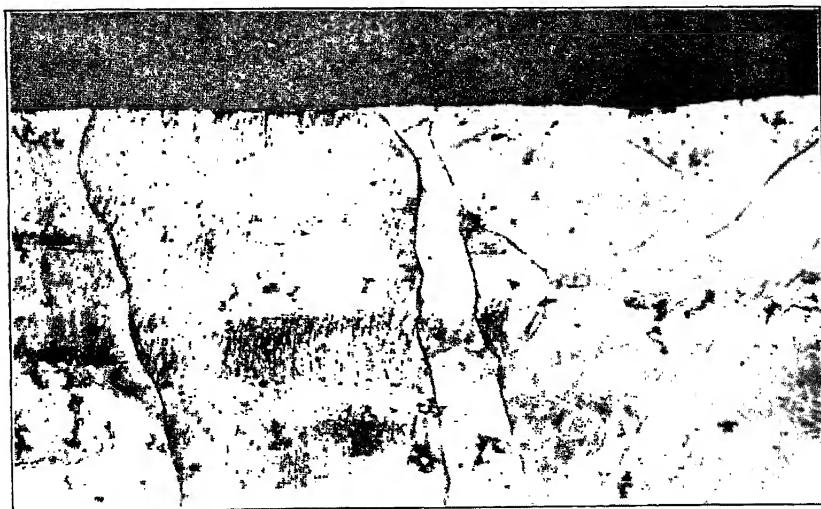


FIG. 92.

Microstructure of surface layer of ingot shown in Fig. 90. $\times 100$.

PLATE LIII.

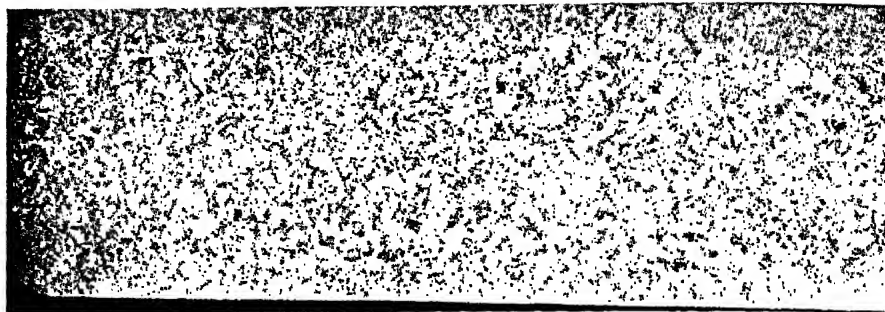


FIG. 93.

Structure of 91/9 aluminium bronze poured at low temperature. Actual size.



FIG. 94.

Structure of 91/9 aluminium bronze poured at high temperature.
Actual size.

structure obtained under such conditions is shown in Fig. 93. With a higher temperature the columnar structure illustrated in Fig. 94 is obtained; this is from an ingot which cracked badly on rolling.

Apart from the use of the Durville process for copper-rich alloys containing added aluminium, the method is entirely satisfactory for the casting of pure aluminium and aluminium-rich alloys. Pure aluminium has been cast successfully in this way at a temperature of 680°C . (*i.e.* only 20°C . above the freezing point); and a fine structure and high degree of soundness can be readily obtained together with a smooth surface. In view of the similarity in density between metallic aluminium and its oxide, there might be some advantage, when dealing with such material, in so designing the apparatus that the melting furnace can be attached to the mould and take the place of the ladle. By this means the aluminium alloy could be melted and cast without the risk of introducing suspended oxide during transfer from the melting furnace to the Durville ladle.

While the advantages of the rotatory method are particularly marked in connection with the casting of materials containing aluminium, the process is not limited to such alloys. For ingots of pure brass, where it is not desired to add any other constituent to the melt to give a stronger oxide skin, the method can be used satisfactorily by the adoption of some auxiliary means for avoiding the formation of oxides on the metal surface. As in the modified crucible casting method (described on p. 129), the use of an external coal-gas flame, applied so as to fill the space between the mouth of the ladle and the mould, gives satisfactory results. Alternatively, oxidation may be avoided by the use of a suitable flux, and ingots of high quality can be regularly produced by such methods. Ingots of pure brasses cast by the Durville process are not only of high surface quality but are completely free from subsurface porosity and from the minute cavities which commonly occur in ordinary top-cast ingots owing to the injection of gases. The finally rolled material is consequently free not only from major defects such as spills and large blisters, but also from the small type of blister previously referred to (p. 30) as a common trouble in the manufacture of cross-rolled brass sheet. It is in fact

probable that for the casting of strip ingots of pure brasses of the highest possible quality the Durville process is the most satisfactory method at present available. Similarly, there is no insuperable difficulty in the adaptation of the method for the casting of other aluminium-free materials, such as nickel-brasses, special brasses, bronzes, zinc, *et cetera*.

The commercial application of the Durville process has in this country been limited almost entirely to the manufacture of strip ingots and solid billets for tube manufacture. Hollow tubular castings of comparatively short length can be made without difficulty, but the use of the method in such directions depends largely on the dimensions of the casting in question. Long tubular castings of thin cross-section are not suitable for rotatory casting, owing to the difficulties of running the metal sufficiently smoothly and slowly in a constricted space, the high temperature necessary for long thin castings, and the disturbance occasioned by cores. Thus the application of the method and design of the apparatus for a particular shape of casting or composition of material is a matter for consideration according to the circumstances in each particular case. In general, however, the following points should be borne in mind in the construction of a plant for commercial use.

1. The lower edge of the ladle and that of the mould must be in a straight line, free from steps or irregularities which might cause a stream of metal to run into the mould.
2. The length of ladle and runner together should not be greatly less than that of the mould. This is necessary to avoid a fall in the height of the liquid at the mouth of the mould during the early stages of pouring. A salamander crucible forms a satisfactory material for lining the ladle; it does not flake readily and any residual skull of metal is easily removed.
3. To ensure rapid working with one machine a number of moulds should be available. This involves the provision of some method for rapid exchange of moulds without difficulty or undue delay.
4. The casting temperature should be as low as possible. For high melting-point alloys, 50° C. above the liquidus

should allow a satisfactory margin, and this can be reduced considerably for metals such as aluminium or zinc.

5. Skimming of the surface of the metal in the ladle must be done carefully to avoid the carrying of dross into the mould.
6. Rotation should be rapid and smooth.
7. While no mould-dressing is essential, a thin coating of lamp-black is desirable to reduce the initial chill of the ingot surface.
8. In casting alloys which do not form a strong oxide film on the surface, the metal should be protected from oxidation.
9. Piping, which is usually small when the casting temperature is sufficiently low, can be avoided completely by the use of a fire-clay dozzle.

The Erical Process.

In the brief review of new developments in casting methods (p. 13), reference has already been made to a new type of water-cooled mould devised by Erichsen.* In this mould, which is known as the "Erical," the cooling plates or faces consist of a nickel-iron alloy of exceptionally low heat conductivity,† which possesses in addition a low coefficient of thermal expansion. It is claimed that as a consequence of this the rate of solidification of the ingot is slower than in an ordinary mould.

During solidification the inner faces of the plates are at a higher temperature than the outer faces. The resulting unequal expansion causes the plates to bend so that the inner surface of the mould tends to become convex, thereby exerting such a strong pressure on the ingot during solidification that liquid metal from the central portion is forced upwards. This is referred to as the "Erical effect." In thin ingots (up to 2 inches thick) the liquid metal is exuded in a sufficient quantity to compensate completely for shrinkage, and piping is avoided without any need for feeding; in some cases the ingot is actually increased in length by this effect. With thicker ingots (*i.e.*

* British Patent No. 358,697 (1931).

† The thermal conductivity of the alloy is approximately one-fifth that of iron or one-thirty-fifth that of copper.

2 to 4 inches) a small amount of piping may occur with certain casting conditions.

The "Erical" effect is stated to occur similarly in all brasses, bronzes and nickel-silvers. The bending of the plates on sudden heating from one side is clearly the direct result of the presence of a steep temperature gradient in the plate and originates in the same manner as the cracking of cast-iron moulds or the opening out of such moulds which may be observed when the clamps are not tightly fixed.

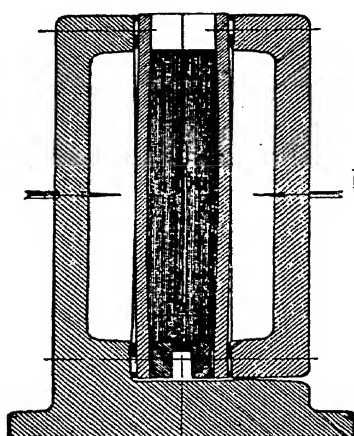


FIG. 95.

Construction of Erical mould.
Position of plates prior to casting.

PRESSURE EXERTED
BY THE ERICAL PLATES
DUE TO THEIR
EXPANSION.

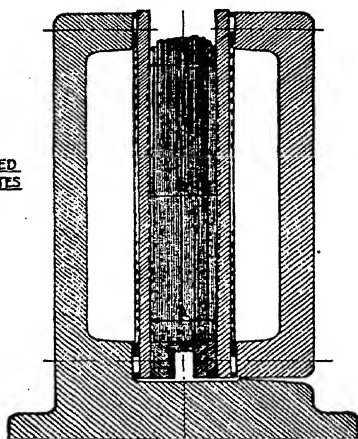


FIG. 96.

Construction of Erical mould.
Solidification of ingot under
pressure of mould wall.

The pressure effect in the "Erical" mould is further claimed to give uniformly fine ingot structure, a favourable factor for subsequent working. The rolling properties of the cast material and the workability of the finished sheet are stated to be considerably improved.

In order that the ingots should not be affected by cracks in the middle region due to contraction effects on the portion restricted by the inward pressure of the plates during solidification, the plates are initially given a reverse bend (toward the water side) by means of packing screws. The stresses produced during casting are thus roughly compensated and the plates

PLATE LIV.

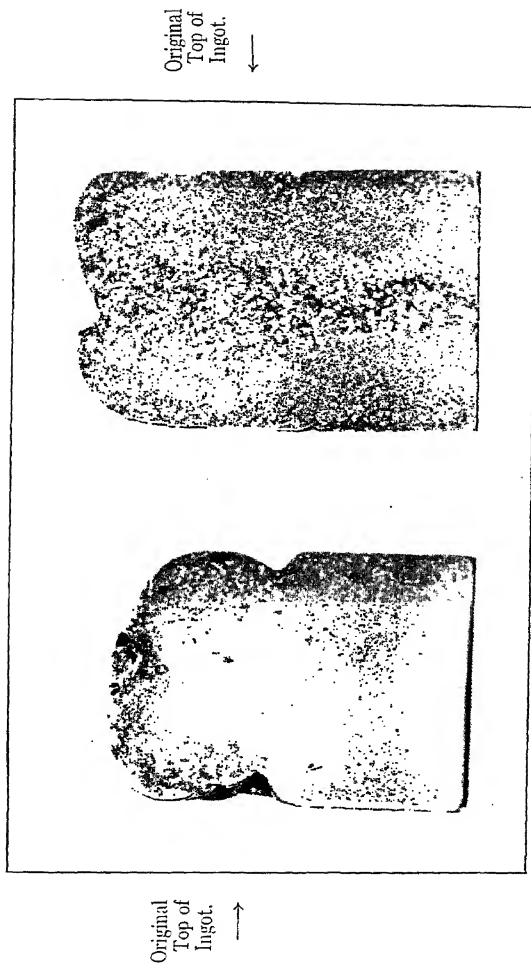


FIG. 97.

Sections through top of Erial ingot, showing exudation of material during solidification.
Actual size.

Edge of
Ingot.

Centre
Line of
Ingot.

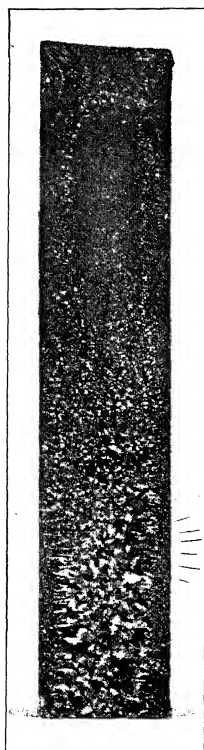


FIG. 98.

Cross section of Erial brass ingot

Edge of
Ingot.

Centre
Line of
Ingot.

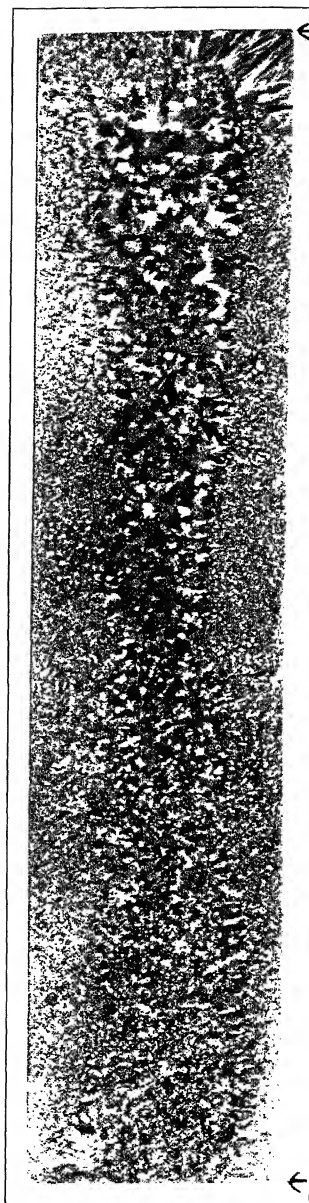


FIG. 99.

Section of commercial 70/30 brass ingot cast
in molten fused water cooled mould

become approximately flat, the ingot having at most only a slight degree of concavity in the centre.

The construction of the mould is shown in Figs. 95 and 96, the necessary flexibility of the joint between the plates and water jacket being obtained by means of rubber packing. The mould is used in the vertical position, the metal being poured through a tundish. If the streams impinge on the mould surface there may be danger of adhesion, and it is therefore important that the tundish should be positioned accurately. A flaming mould-dressing (oil base) is applied on a thin under-coating of clay. Cleanliness of the "Erical" plates is of great importance and special methods of cleaning are recommended.

The top portion of a typical ingot ($22 \times 19.5 \times 1.5$ inches) cast by the "Erical" process, showing the characteristic exudation of material from the centre, is illustrated in Fig. 97. The macrostructure of a 65/35 brass ingot cast by the "Erical" process is illustrated in Fig. 98. This is typical of brass ingots cast in moulds coated with a volatile dressing. The crystals on the whole are slightly finer than those of ingots cast by ordinary methods, but larger crystals are found in some portions of the ingot, particularly the lower half. Comparison of the structure of the ingot (Fig. 98) with that of a commercial brass ingot (Fig. 99) cast in a copper-faced water-cooled mould suggests that the fine structure is generally attributable to the influence of the volatile coating rather than to any specific effect of the mould material.

As regards internal soundness, "Erical" ingots are, as would be expected, somewhat superior to those obtained in ordinary practice, but density determinations show the presence of cavities to an amount of the order of 0.35 per cent. by volume. These may be due to the entrapping of injected gases, a type of defect which would not necessarily be eliminated by pressure during solidification.

In view of the claims that "Erical" brass is of finer grain than brass cast by ordinary methods and is therefore a more suitable material for cold working, the authors have made a comparison of the mechanical properties of strip rolled from an "Erical" ingot with those of material of similar copper content made in a cast-iron mould by the crucible casting process. Strip rolled from the annealed "Erical" ingot was found to be

inferior in the as-rolled condition (without subsequent annealing) in both strength and ductility to the ordinary commercial product; but on further rolling with intermediate annealings, the "Erical" material at each stage showed a higher maximum stress and a lower percentage elongation. The hardness values followed those of maximum strength, while cupping tests (Erichsen and Guillery) gave better results than the ordinary commercial product. The microstructure of the cold worked and annealed "Erical" brass was at all stages finer than that of the ordinary material. Typical structures of strips annealed at 650° C. after reduction by 60 per cent. in two stages, are shown in Fig. 100. This finer grain of the "Erical" material, which provides the explanation of such differences in mechanical properties as were observed, was considered to be due to the chemical composition, considerably more phosphorus and iron being present than in the commercial ingot. The differences in structure and properties could not, therefore, be attributed to the method of casting.

It may be generally concluded that, from the point of view of quality of the cast ingot, the "Erical" process is beneficial only in the direction of improving soundness as a result of the compression exerted during solidification. It is improbable on theoretical grounds that a casting method can produce improvement in the working properties of strip, which is subjected to several stages of rolling with intermediate annealings, and this is borne out by examination.

While the "Erical" mould is novel in principle and of considerable metallurgical interest, especially in connection with the casting of alloys liable to segregation effects, it possesses disadvantages in manipulation, and in this respect may compare somewhat unfavourably with the water-cooled copper moulds used for brass casting. While no definite information as to the life of the special alloy plates is available, it appears reasonable to assume that the severe stresses set up within the "Erical" plates would lead either to cumulative warping or to surface cracking. Provision is made in design for the periodical turning of the plates. This would diminish the tendency to distortion, but to be effective the changes would need to be frequent, and might involve difficulty in the foundry.



(a)



(b)

FIG. 100.

Microstructure of 65/35 brass strip annealed at $650^{\circ}\text{C}.$, after reduction by 70 per cent. in two stages. $\times 50$.

- (a) Erical brass.
- (b) Ordinary crucible cast brass.

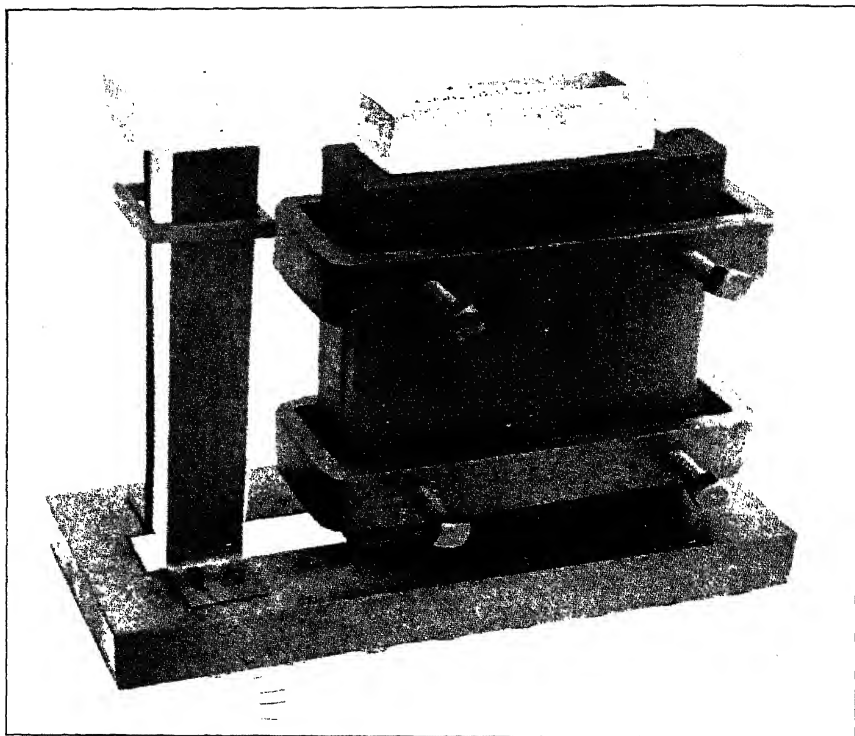


FIG. 101.
Experimental apparatus for bottom casting.

Bottom Casting.

The process of bottom casting is well known in the steel manufacturing industry, where it has become increasingly used. It has the advantages of giving ingots of good surface and subsurface soundness and of dealing conveniently with large volumes of steel. The moulds, a number of which are filled at the same time, are open at both ends and stand on a bottom plate in which are fire-clay tubes or runners leading from each mould to a central column or "trumpet" where the molten metal enters.

Application of the bottom-casting process to brass casting is of interest in view of the desirability of avoiding the turbulence which is characteristic of the ordinary top-casting method. When ingots of 70/30 brass are bottom-cast in air the formation of a surface oxide film on the metal rising in the mould results in defective regions on the ingot surface. By casting with the mould leaning at a slight angle the oxide film can be made to adhere to one face only, and an ingot is produced having one face almost perfectly smooth and free from defects, while the opposite face is uniformly covered with oxide skin folded along curves produced by the flow of the metal. With a volatile mould coating the ingots produced are similar in character to industrial top-poured ingots of the highest quality.

Both the gas-flame method and the use of phosphorus addition can be successfully applied in adapting the bottom-casting process for use in the manufacture of brass ingots.

In the bottom casting of experimental $12 \times 6 \times 1.5$ inches brass ingots, the cast-iron mould is of the usual type but open at both ends, and rests on a steel plate in which is a slot containing a moulded fire-clay bottom runner. A vertical fire-clay runner is luted to the end of the bottom runner. The metal, poured from a bottom-pouring ladle, passes down the vertical runner, along the bottom runner, and enters the mould through a row of holes. A hot dozzle may be used to facilitate feeding, and the mould is coated with soot. Fig. 101 illustrates an improvised apparatus of the type described.

The application of the gas-flame method (used for pure brass) necessitates the use of two burners, one over the vertical runner and another over the top of the mould, for the complete

protection of the molten metal from oxidation. No flame protection need be used when phosphorus is added to the brass. The surface quality of the bottom-cast ingots, whether of pure 70/30 brass or phosphorus brass, is definitely superior to that of corresponding top-cast ingots and there is normally little difference in surface quality between phosphorus brass bottom-cast in air and 70/30 brass similarly cast with a gas flame. Slight rippling at the edges is liable to occur near the top of the ingot, but this feature is of no disadvantage. Bottom-cast ingots give rolled strip of high quality, and numerous tests comprising the manufacture of drawn articles have shown no difference in working qualities between top-cast and bottom-cast ingots of the same alloys.

Where the smoothest possible ingot surface is required, the bottom-casting process appears to have distinct advantages, especially in connection with the use of phosphorus as an addition to brass. In crucible casting, where the individual melt is small, the manipulative difficulties and the large amount of scrap due to the runner render bottom casting uneconomical, but in view of the progressive increase in size of melting units which is a feature of recent advances in works methods, it appears possible that eventually the bottom-casting process may become a useful feature of non-ferrous industrial practice.

APPENDIX A

THE MANUFACTURE OF BRASS STRIP AND SHEET

Development of Rolling Mills—Modern Rolling Mills—Annealing Practice—Pickling—Cold Rolling of 70/30 Brass Ingots—Hot Rolling.

IN very early times ingots of copper and brass were made into plates by hand hammering, and copper sheets of primitive manufacture have been found by archæologists among early native tools and ornaments. The method of hammering sheet persisted for many years, and with the advance of mechanical knowledge was developed by the use of power-driven hammers, usually driven by water. These constituted the only type of plant used for making sheet and strip until about the end of the seventeenth century, and in many works persisted much later.

The method of forming sheets by mechanical hammering was known as the "battery" process, and the term "battery" still occurs in the names of some individual works. The battery process was introduced into this country from Germany in the sixteenth century, the sole right for its use in the fabrication of copper and brass being granted to the Society of Mineral and Battery Works in 1568. The hammers used were of various weights, some of them weighing as much as 500 lb., and were operated by water power. The cast ingots were first hammered into plates of suitable thickness. These were cut to shapes and fashioned by further hammering into the final form required.

While the battery process continued to be used in the manufacture of finished articles such as vessels of various kinds, rolling mills, when once introduced, became increasingly used for production of the initial sheet. The first rolling mill was devised early in the sixteenth century by a goldsmith named Marx Schwab in Augsburg.¹ The earliest mills were turned by hand and the method made little progress until the use of

water power became more general. Reversing mills were developed in England for rolling sheet lead, the first probably being that built in Deptford in 1670 by Thomas Hale.² In 1697 the first copper-rolling mill was established in England at Dockwra Copper Co., though it was not until Watt developed the steam engine that rolling mills came generally to supersede battery works for this purpose. The original Dockwra Mill is probably similar to that described by Swedenborg,³ who gives probably the fullest available account of the early operation of rolling. An extract is given below :—

“ When the brass has been poured out and formed into tolerably thick plate it is then cut . . . into smaller plates 2 inches broad and $4\frac{1}{2}$ ft. long, these . . . are flattened out between two iron rollers. The rollers are about 6 inches in diameter and are turned by two water wheels, but before this brass is thoroughly flattened out by the rollers it is made red hot. . . . These rollers are made of wrought iron and are made round by steel instruments; and they cannot be in use longer than 2, 3 or 4 days, without requiring to be again repaired at the turning lathe.”

The use of such mills in Sweden was largely due to Christopher Polhem (1661–1751), who took a prominent part in the development of early mechanical plant. Polhem devised methods of hardening steel rolls and of grinding them subsequent to hardening, but his most notable invention was the four-high mill using “ two slender wrought-iron rolls between two thick cast-iron rolls which prevent the slender rolls from yielding.” ⁴

The development of modern rolling mills from these early beginnings has been along varied lines, and a wide variety of plant is available at the present time for different types of work. In the cold rolling of 70/30 brass to strip and sheet the simple two-high mill is most generally used, the diameter of the rolls decreasing with the gauge of the metal rolled. For the production of sheet the rolls are long, and the diameter is necessarily increased to give adequate strength. Small diameter rolls, however, bite into the metal with greater ease and require less power as a result of the smaller surface of roll in contact with the work. Small diameter rolls can be used, as Polhem showed, provided the necessary support is given by auxiliary large diameter “ backing up ” rolls. A variety of

types of mill embodying this principle are available, and their use in non-ferrous metal sheet manufacture is increasing. The simplest form is the four-high mill, which consists of two working rolls of small diameter backed up by large diameter supporting rolls. A modification of this type is the six-roll or cluster mill where each of the small diameter working rolls has two supporting rolls. In this design bearings for the working rolls can be dispensed with and the working rolls may be smaller in diameter than in a four-high mill used for the same work. The cluster mill is now used in the manufacture of cold-rolled brass sheet and strip, being most serviceable in the production of sheet of thicknesses below 0.15 inch. In brass of this thickness comparatively heavy reductions (50 per cent. or more in thickness) can be made in a single pass.

In order to permit the use of even smaller diameter working rolls than is possible in a six-roll cluster mill, a twelve-roll mill has been devised ⁵ in which the four backing-up rolls are of insufficient diameter to take the full load and are accordingly themselves supported by six large rolls which stand well clear of the work.

Another interesting type of backed-up mill (Lauth ⁶) has three rolls of which the top and bottom are of large diameter and driven, while the middle roll of smaller diameter is free. The middle roll can be brought in contact with either the top or bottom rolls, and the mill may thus be used in the same way as a three-high mill. The Lauth mill has advantages for hot rolling, but in cold rolling the differences in diameter between the two working rolls is liable to cause internal stresses in the strip, and loss of power.

In some instances, two-high mills are arranged in tandem in order to expedite output. The reduction and speed of each set of rolls are so adjusted that the strip is taken continuously through the series, undergoing the maximum reduction possible before annealing is necessary. No new principle is involved in this, and designs of mills working on this principle were patented by Hazeldine in 1798.⁷ Continuous two-high mills are not frequently used for the cold rolling of strip, but find more general application in wire rolling, where the difficulty of maintaining a uniform tension over a considerable width does not arise.

Finally, mention should be made of a later type of strip mill (Steckel⁸), which has been devised on an entirely new principle. In this mill the rolls are not driven but run freely while the strip is pulled through by power-driven reels. The working rolls of small diameter are supported by large backing-up rolls. It is claimed that the material rolled can be reduced in thickness to a greater extent than in the ordinary type of mill without requiring annealing. Use of the Steckel mill has been largely confined to steel practice, though it is stated to be equally suitable for non-ferrous metals.

The necessity for annealing metals after a certain amount of cold working was recognised in early manufacture, and illustrations of the battery process suggest that the annealing method first employed was a simple heating in a fire similar to that of the blacksmith. Furnaces were later developed in which the plates of rolled metal were stacked on a hearth and heated by exposing them directly to the flames of a coal fire. Such methods, however, not only exposed the material to deleterious products of combustion, but resulted in uneven heating. In modern practice closed furnaces of the muffle type are employed in which a charge of rolled strips weighing from about 0.5 to 3 tons in weight can be heated to a uniform temperature without exposure to the furnace gases. Electrically heated muffles have shown rapid development, and in modern practice electric furnaces for continuous annealing of rolled strip are receiving consideration.

In all such furnaces a certain amount of surface oxidation is inevitable and in the early stages of rolling, while this is reduced to a minimum, no special methods are employed to eliminate it entirely. For the final annealing a variety of methods, such as the use of water-sealed muffles, pot annealing, etc., are employed to achieve this object. The use of special "bright annealing" methods is increasing, and the difficulties in the treatment of brass by these methods have been largely overcome.

Where the atmosphere of the annealing furnace is oxidising, methods of cleaning are employed for removing the superficial oxide scale. In the early stages of rolling the brass is treated by pickling in a sulphuric acid solution containing usually about 10 per cent. by volume of sulphuric acid. Where the scale

contains cuprous oxide, this reacts with sulphuric acid to give a thin deposit of metallic copper on the surface of the brass, and the resulting "red stain" spoils the appearance of the finished sheet. After the final annealing, therefore, pickling solutions having an oxidising action are sometimes used, such as mixtures of nitric and sulphuric acids, chromic acid, etc. Such solutions have a tendency to "etch" the metal, showing the crystal structure, and "close" annealing methods which prevent any considerable formation of cuprous oxide in the scale are therefore preferable.

Although rolling varies in detail, the following may be regarded as average cold-rolling practice in the production of strip from 70/30 brass ingots. The ingot is first cropped until the cut face shows freedom from piping defects. The normal thickness of ingot is about 1 to 1.5 inches, and in the first stage of rolling or "breaking down" stage, heavy reductions are necessary to avoid internal stresses in the strip. A reduction of about 50 per cent. is generally given in two or three passes on mills of large roll diameter (24 inches) working at about 100 ft. per minute. The material is then annealed at a temperature of about 600–650° C. Scalping, *i.e.* machining or scraping of the ingot surface to remove superficial casting defects, is usually carried out at this stage, as the removal of the camber from the original ingot facilitates the scalping operations. Scalping, however, is not universal, and where used it is accomplished by a variety of methods. In some instances the surface layers are removed by machining the ingot faces with rotary cutters before rolling.

After annealing, the strip is cleaned in a sulphuric acid dip and is then ready for its next stage of rolling, which again usually involves 50 per cent. reduction in thickness. The material is taken through successive operations of about 50 per cent. reduction and annealing until the finishing stage, where the amount of reduction depends on the final "temper" required in the finished sheet. With the type of two-high mill still widely used, the number of passes necessary to secure a 50 per cent. reduction in thickness at each stage increases as the material becomes thinner, five or six passes being necessary for strip of the order of 0.03–0.06 inch thick. With modern powerful high-speed two-high mills, however, this

increased number of passes is not necessary, and heavy reductions can be made in a single pass. Speeds of rolling for intermediate and finishing mills are of the order of 150–200 ft. per minute, and the modern tendency is to increase this, for finishing thin strip, up to 250–400 ft. per minute.

Cold rolling by the method described is suitable for brasses of all copper contents higher than about 61 per cent. Below 65 per cent. copper a decrease in the copper content produces an increase in hardness, and in the cold rolling of a brass containing 62 per cent. copper, the heavy passes used for 70/30 brass are not practicable. Brasses of still lower copper content (60–56 per cent.) harden so rapidly on working that cold rolling is not economically possible.

The hot rolling of brass was first carried out by Muntz in 1832 on brass of about 60 per cent. copper content. Alloys based on this composition (61–57 per cent. copper) are readily workable at all temperatures from about 550° to 800° C. and have an extensive field of use. 70/30 brass, however, does not undergo the phase change which in 60/40 brass confers easy workability at high temperatures; although increased in ductility at temperatures of about 650°–850° C., it is much less ductile than 60/40 brass at these temperatures and is liable to become brittle in the neighbourhood of 500° C. Hot rolling of 70/30 brass consequently presents difficulty in that the working must be completed in a short time, and purity, especially freedom from lead, is essential. In recent years the commercial hot rolling of 70/30 brass has been satisfactorily established, a development of wide importance in view of the greater thickness of ingot which can be handled.

The type of mill used for hot rolling is essentially one capable of rapid reduction of the ingot thickness, and for this purpose either a three-high mill with tilting tables on both sides, or a two-high reversing mill is suitable. Tandem continuous mills are also convenient for hot rolling. A 70/30 brass ingot 4 inches in thickness can be rolled to strip about 0.2 inch thick in one heat, assuming an initial rolling temperature of about 850° C. Further reduction and the finishing passes are then carried out in the cold. The rapidity with which the hot-rolling mill enables the material to be handled results in a considerable economy in rolling costs, but a complete change-over from cold

rolling practice to hot rolling involves both the installation of new rolling plant and a modification of casting methods to produce the large ingots which are necessary to secure the maximum benefit. Hot rolling of 70/30 brass is consequently being developed first in the larger works, and it seems probable that the older methods of manufacture of strip by cold rolling from thin ingots is likely to account for a considerable output of brass strip for many years to come.

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- ⁵ Brit. Pats. 354,234 and 363,970. W. Rohn, *Metallurgia*, 1932, 7, 47.
- ⁶ Brit. Pat. 2813, 1862.
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- ⁸ *Iron Age*, 1932, 129, 168 and 214.

APPENDIX B

CONSTITUTION AND DENSITY OF THE INDUSTRIAL BRASSES

Crystal Structure of Metals and Alloys—The Equilibrium Diagram—
Structure of Typical Brasses—The Density of the Brasses.

FOR an understanding of the factors involved in the casting, working and annealing of metals, some knowledge of constitution and structure is necessary. All solid metals and alloys consist of small crystalline grains, and in a pure metal these individual crystals are uniform in composition. Fig. 102 shows the microstructure of pure lead and is representative of the structure of most pure metals, which, apart from colour differences, show the same general microscopic features.* Each "crystal grain" is built up on a regular geometrical pattern or lattice, but is prevented from assuming a geometrically perfect contour by the mutual interference of adjacent grains during solidification. The individual crystals commence their growth with their axes in different directions, and this variation in "orientation" prevents them from growing together into one uniform crystal.

The crystallisation of alloys is more complicated and variable, but it is only necessary to consider here the case of alloys of two metals (such as copper and zinc) which in the molten state can be mixed together in all proportions to form a homogeneous solution somewhat analogous to a solution of water and alcohol. When such a liquid metallic solution solidifies the simplest cases vary between two extremes: (1) the state of solution is resolved, with the formation of a mixture of crystals

* The surface of a metal specimen for micro-examination is ground flat and polished, and finally etched by means of a suitable reagent to reveal the structure. In many alloys, different types of crystal are distinguished by their form and colour, and in pure metals differences in orientation of the crystals affect the direction in which light is reflected, adjacent crystals varying in depth of tone.

of the two elements, and (2) the state of solution is maintained during solidification, resulting in crystals of individually uni-

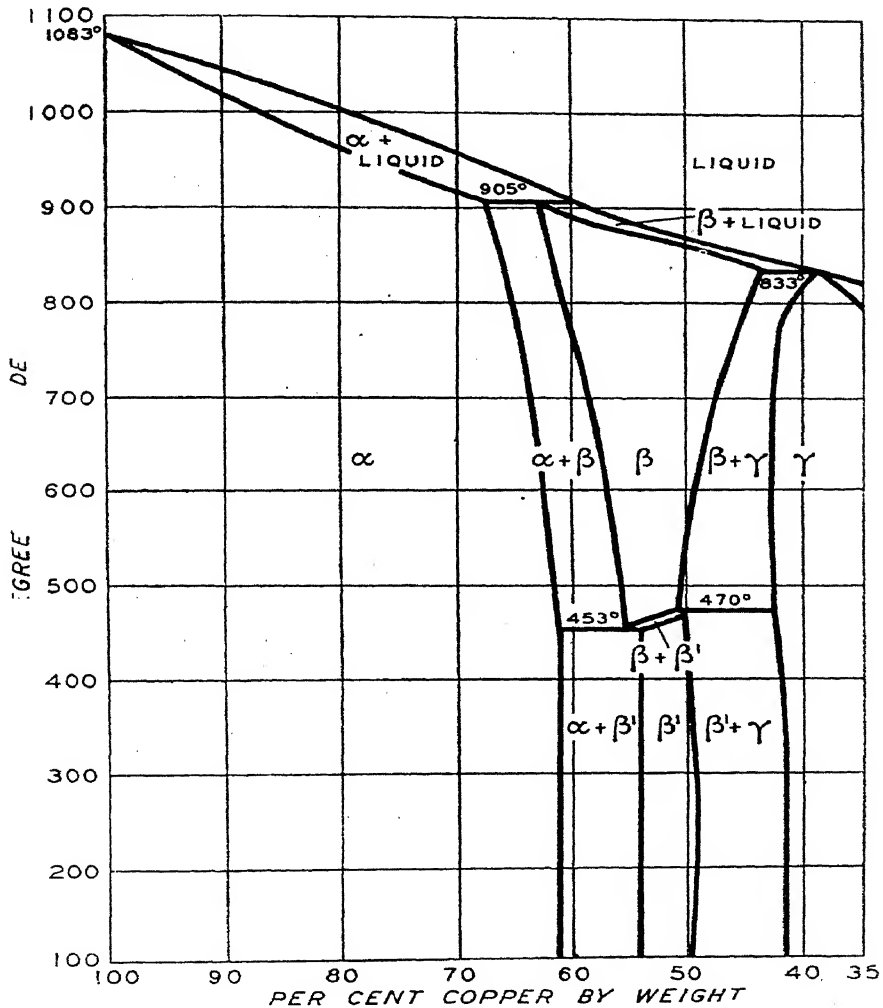


FIG. 103.

Equilibrium diagram of copper-zinc alloys : 35 to 100 per cent. copper.

form composition containing both metals ("solid solutions"). In the majority of industrial binary alloys, a single solution is

formed only over a limited range of composition. Within other ranges additional constituents are formed which may be further solid solutions or crystals of entirely different character, such as metallic compounds. The mechanical properties of an alloy vary considerably with differences in constitution. Pure metals and solid solutions are in general, though not invariably, ductile, while compounds are characteristically hard and brittle.

As an alloy is cooled after solidification, a change in structure may occur owing to variation of solid solubility with temperature, and such changes are the basis of the processes known as heat-treatment.

It is customary to represent the constitution of an alloy system as an "equilibrium diagram" (in which composition is plotted horizontally and temperature vertically) which indicates the limits of composition and temperature of each constituent or "phase." The diagram for alloys of copper and zinc containing up to 65 per cent. zinc is shown in Fig. 103.^{1, 2, 3} This diagram represents conditions of equilibrium, which at low temperatures can only be obtained after extremely prolonged annealing. The practical significance of this in connection with the copper-zinc alloys will be referred to later.

A study of the diagram shows that when zinc is added to copper, the commencement of freezing (liquidus) is depressed continuously from 1083° C. for pure copper to 833° C. for an alloy containing 60 per cent. of zinc. Unlike pure metals, which have a single freezing point, solid solution alloys solidify over a range of temperature, and a 70/30 copper-zinc alloy, which commences to freeze at 960° C. (liquidus), becomes completely solid at 910° C. (solidus). In solid solution alloys freezing in this way, the crystals first separating are different in composition from that of the actual liquid, and there is a tendency for each crystal to be graded in composition from centre to boundary. If solidification is sufficiently slow this difference disappears by diffusion, but in chill cast 70/30 brass the microstructure shows each grain to be richer in copper in the centre and poorer at the edges, an effect known as "coring" and illustrated in Fig. 104. Annealing permits diffusion and results in a uniformity of composition within the grains, forming a structure essentially similar to that of a pure metal.



FIG. 102.
Pure Lead. Extruded. $\times 20$.



FIG. 104.
70/30 brass, as cast. $\times 50$.



FIG. 105.
60/40 brass, as cast. $\times 100$.



FIG. 106.
65/35 brass, as cast. $\times 50$.

The alpha (α) * solid solution of the copper-zinc system, of which 70/30 brass is typical, is capable of holding 32.5 per cent. zinc at the "solidus" temperature of 905° C. and 39 per cent. zinc at atmospheric temperatures. If more than these amounts of zinc are present at the respective temperatures a second constituent appears, designated the beta (β) solid solution.

Over the range of 32.5 per cent. zinc to 37 per cent. zinc at 900° C. and 39 per cent. zinc to 46 per cent. zinc at ordinary temperatures, the alloys consist of a mixture of the crystals of the alpha solution and those of the beta solution, while beyond these percentages of zinc the alloy is entirely beta up to 50 per cent. zinc, which can be taken as the limit of compositions which are useful commercially. Fig. 105 shows the structure of 60/40 brass as cast, a considerable amount of beta phase being present. Beyond 50 per cent. zinc a third constituent appears as shown on the diagram, namely the gamma (γ) constituent, which is extremely hard and brittle and renders the alloy useless for working purposes.

The effect of these variations in structure on the mechanical properties of the alloys is considerable. As the zinc is increased up to the maximum content which is soluble in the alpha phase, the colour of the alloy changes from the red of copper to the yellow of 65/35 brass. The tensile strength and hardness of the alloys increase with the zinc content, while the ductility rises to a maximum at about 70 per cent. copper and then begins to fall before the composition reaches the limit of the alpha solution. The beta solution is much harder and less ductile than the alpha solution at ordinary temperatures, and with its appearance the colour of the alloy again becomes darker.

Alloys consisting only of alpha are particularly suitable for cold working and may be rolled hot or cold. The beta solution can be worked readily at high temperatures but will stand only a small amount of deformation in the cold. Alloys consisting of mixtures of these two phases have intermediate properties and can undergo both cold and hot working to an intermediate extent.

This brief general discussion of the equilibrium diagram has.

* Greek letters are used in general in alloy systems to distinguish between different solid solutions, as their use has been found more convenient than the giving of specific names to the large number of different types of solid solutions known.

been included to give some impression of the fundamental properties of the alloys of different compositions, but it must be emphasised that conditions of equilibrium as represented by the diagram are not always obtained in ordinary practice. For example, in the chill casting of a 68 per cent. copper brass, the rapidly solidified alpha crystals do not take into solution the maximum amount of zinc, and some beta constituent is formed. In 65/35 brass the effect is pronounced (Fig. 106) and a considerable amount of beta phase is present, but on annealing at 700° C. the structure of the alloy becomes uniform alpha. Similarly, in a 62 per cent. brass, which according to the diagram should solidify as beta and change to alpha gradually on further cooling, the rate of diffusion is slow, and the alloy shows residual beta even after prolonged annealing. In brief, the rapid rates of cooling and comparatively short times of annealing given in ordinary practice have the effect of displacing the alpha and beta phase boundaries towards the copper side of the diagram. The limit of solubility of zinc in the alpha phase which is obtained in ordinary practice is of the order of 2 per cent. less than that shown in the diagram, but may vary considerably according to the type of treatment.

Deformation of 70/30 brass in the cold, as by rolling, distorts the crystal grains as shown in Fig. 107, producing increased hardness or resistance to further distortion. On reheating such cold-worked metal a temperature is reached at which the distortion of the cold-worked crystals is relieved by recrystallisation, new crystals being formed which are free from strain. The actual temperature of recrystallisation depends upon the amount of cold work, but may be as low as 350° C., though the growth of new grains is slow at such a low temperature. Fig. 108 shows the fine-grained recrystallised structure produced by annealing for three hours at 350° C. a 70/30 brass which had been reduced 50 per cent. in thickness by cold rolling. As the annealing temperature is raised the rate of crystal growth increases rapidly, and Figs. 109, 110 and 111 show the structures obtained by annealing for half an hour at 500°, 600° and 800° C. respectively. The marked grain growth obtained by annealing at too high a temperature is unsatisfactory from the point of view of response to further working, and the structure shown in Fig. 110 is typical of satisfactory commercial practice.



FIG. 107.—70/30 brass, cold rolled. $\times 100$.



FIG. 108.—70/30 brass, cold rolled, annealed 350°C. for three hours. $\times 100$.



FIG. 109.—70/30 brass, cold rolled, annealed 500°C. for half-hour. $\times 100$.

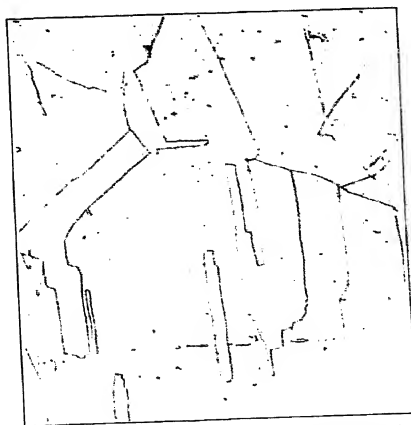
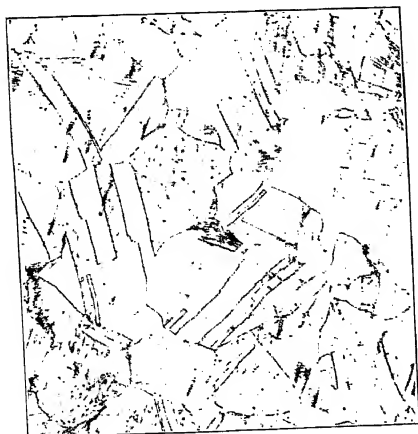


FIG. 110.—70/30 brass, cold rolled, annealed 500°C. for half-hour. $\times 100$.

From a scientific point of view a knowledge of true equilibrium conditions is important, and as a basis for interpretation of observations on the brasses, the authors have carried out the determination of the maximum solubility of zinc in the α solution.²

The equilibrium diagrams previously available gave a limit of solubility of zinc in the α solution as about 37 per cent., but there was some evidence that this figure was too low. Stead and Stedman⁴ noted that in a Muntz metal, which normally consists of approximately equal parts of α and β , most of the β phase disappeared after heating at 430° C. for three months, and the present authors found that brass containing 62 per cent. copper, after annealing at 500° C. for eight days, was converted to uniform α . It is the slowness of the diffusion of zinc in the α solution at low temperatures that renders true equilibrium so difficult to obtain, but the authors were able, by utilising a method somewhat novel in this application, to obtain equilibrium in an alloy containing 61.1 per cent. copper in about one hour. This method consisted in quenching the alloy in iced brine from a temperature of 820° C. so as to produce a structure which was almost entirely β . This structure was extremely unstable at ordinary temperatures, and on re-annealing the quenched alloy at 450° C. for one hour, equilibrium was rapidly obtained and the alloy converted to a uniform α solution. A series of the alloys was therefore taken containing 61.1, 62.0, 63.2, 64.0, 65.2 and 67.7 per cent. copper. By quenching and re-annealing in the case of the lowest copper content, and by direct annealing of the higher copper alloys, all were obtained in the condition of uniform α . Small samples of each alloy were rapidly quenched from different temperatures and examined for the first appearance of β . For each composition temperatures were determined (at not more than 20° intervals) between which the change α to $\alpha + \beta$ occurred. This enabled the boundary of the α phase to be accurately fixed on the diagram.

A knowledge of the density of alloys, *i.e.*, the weight in grams per cubic centimetre, is of considerable importance in that a quantitative measure of the unsoundness can be obtained provided that the density of sound material of the same composition is accurately known. Determinations on brasses

have been published by Bamford,⁵ but there has been some doubt as to whether the samples used were perfectly sound, and the state of the alloys relative to equilibrium was not defined. The densities of the copper-zinc alloys were therefore redetermined by the authors, with the highest possible accuracy, up to a zinc content of 46 per cent.,⁶ and the values obtained have been used in the calculations of the degree of unsoundness given throughout the present book.

One-inch thick ingots were rolled without annealing to a thickness of 0.6 inch, and from this material cylinders 0.5 inch in diameter and 1.5 inches in length were machined. The rolling applied was insufficient sensibly to affect the density apart from the removal of possible unsoundness, and the densities given in the second column of Table XVI may therefore be taken to indicate the values given by perfectly sound chill castings.

Micro-examination, carried out on a number of sections from each slab, showed a cored α structure in the alloys above 71 per cent. copper. Below that composition β was present, increasing in amount as the copper content decreased.

TABLE XVI
DENSITY OF THE INDUSTRIAL BRASSES AFTER VARIOUS TREATMENTS

Copper Content. %.	Density as Cast and Rolled. Gms./c.c.	Density after Annealing to Equilibrium. Gms./c.c.	Density after Quenching from Cast and Rolled Condition. Gms./c.c.	Quenching Temp. ° C.
99.99	8.933	8.933	—	—
96.5	—	8.886	—	—
90.0	8.795	8.793	—	—
80.2	8.657	8.654	—	—
70.0	8.512	8.500	—	—
65.2	8.457	8.400	8.456	820
62.0	8.410	8.368	8.424	820
59.8	8.382	8.370*	8.397	820
51.7	8.292	8.294	8.298	700

* This specimen was not completely in equilibrium.

The densities were determined by the usual method of weighing in air and water, and the values for density in grams

per c.c. calculated according to the generally accepted formula :

In order to obtain comparative values for the alloys in structural equilibrium, a number of cylinders of each composition were annealed for long periods and the densities redetermined after machining 0.01 inch off the surface. The values obtained for the density after these annealing treatments are given in column 3 of Table XVI. It will be seen that with a copper content of more than 71 per cent. the density of the chill cast alloys is not sensibly changed by annealing. Where the alloys as originally cast contained β which was subsequently converted to a certain extent to α by annealing, this change in structure was accompanied by a fall in density : the maximum difference of 0.06 gram per c.c. occurring at a composition of 65 per cent. copper. The samples in the range 60 per cent. to 65 per cent. copper, which show this decrease in density on annealing, were heated to 820° C. for 45 minutes, quenched and the density redetermined, and this treatment was found to cause an increase in the density to a figure somewhat exceeding that given by the original rolled strip. It would appear, therefore, that for a given composition, increasing the proportion of β phase relative to α causes a rise in density and *vice versa*. The drop in density on annealing which is shown in Table XVI is not, however, entirely due to the phase change, as it was found to be accompanied to a certain extent by the production of porosity. Compression or other mechanical deformation of the annealed samples in this range raised the density to a figure approaching that given by chill cast material. There was, however, a definite marked difference in the range 60 to 70 per cent. copper between the density of sound chill cast or quenched metal, and of sound material in structural equilibrium.

This variation is particularly marked in brasses containing 63 to 65 per cent. copper. Annealing such alloys after rolling produced in several specimens of each composition a drop in density of 0.04–0.06 gm. per c.c. Requenching this annealed material raised the density to a figure higher than the original figure for cast material by 0.01–0.02 gm. per c.c.,

while subsequent annealing again produced a drop in density, which in this case was somewhat less than the drop on the first anneal. The results obtained are summarised in Table XVII and from these, together with the figures given in Table XVI the curve shown in Fig. 112 has been drawn. The true density of annealed brasses is taken as that given by material compressed after annealing.

TABLE XVII
EFFECT OF COMPRESSION IN RAISING THE DENSITY OF
BRASSES ANNEALED AFTER ROLLING

Cop- per Con- tent. %.	Original Condition of Alloy.	Density in Original Condition. Gms./c.c.	Density after Anneal- ing. Gms./c.c.	Density after Compression of Anld. Cylinders. Gms./c.c.	Density after requenching Compressed Cylinders. Gms./c.c.
70.0	Cast and rolled	8.512	8.495	8.508	—
65.2	do.	8.456	8.396	8.445	8.460
64.0	do.	—	8.362	8.432	8.436
64.0	Quenched	8.448	8.429	—	—
63.2	Cast and rolled	—	8.361	8.422	8.433
63.2	Quenched	8.431	8.425	—	—
61.1	Cast and rolled	—	8.366	8.384	—
61.1	Quenched	8.416	8.400	8.399	8.414

This curve represents with a high degree of accuracy the density of sound material over the range of composition in question. The examination of a large number of industrial samples of rolled brass has given values in close agreement, and in no case higher than those indicated in the graph. The difference in density resulting from heat treatment over the range 55 to 65 per cent. copper is of considerable practical interest. Heat treatment involving the quenching of brass from within this range may give rise to internal stresses of considerable magnitude due to the constitutional changes taking place during cooling. An interesting example of this type of failure occurred in a brass fuse-body containing 57.2 per cent. copper.² This had apparently been quenched from a temperature within the β range, and while in the outer region the rate of cooling had been sufficiently rapid to retain a structure consisting

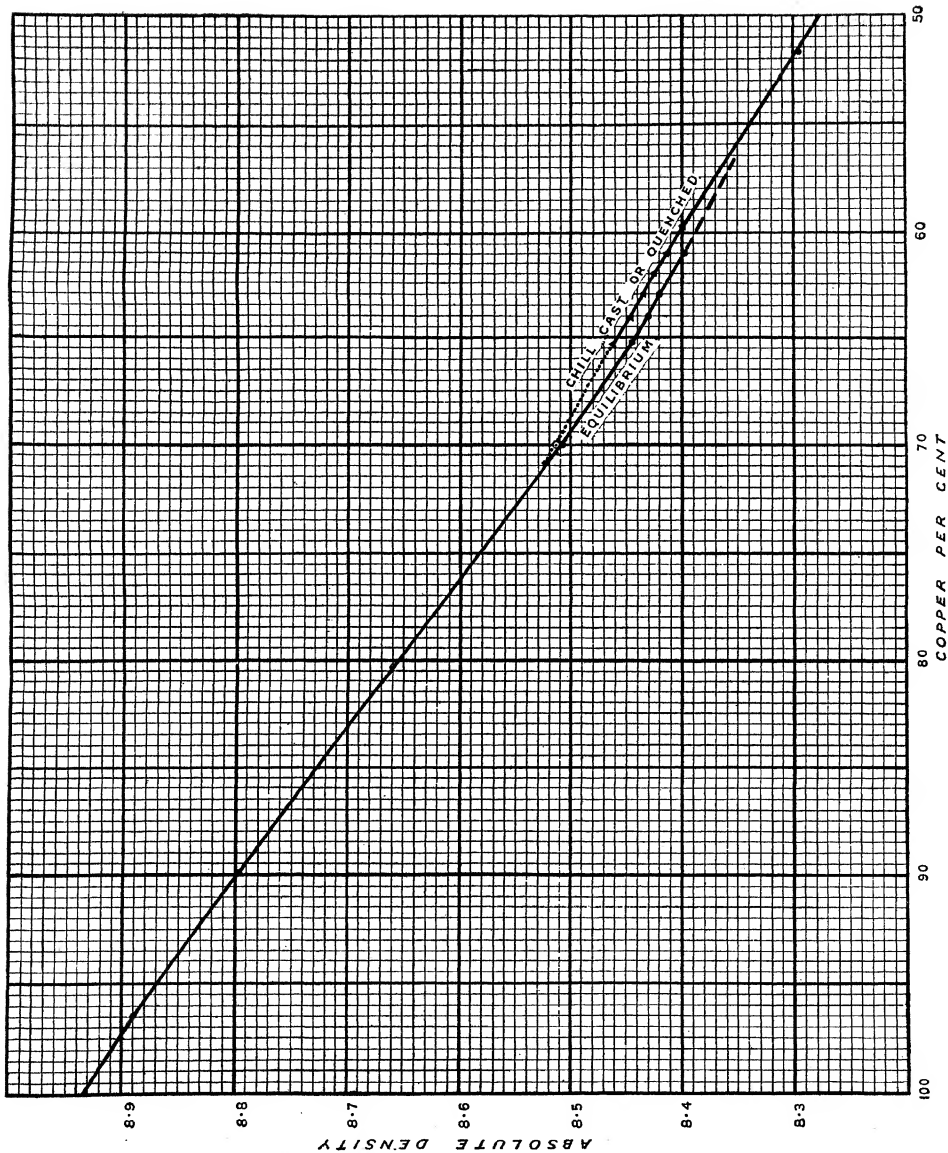


FIG. 112.
Curve of density of the brasses containing 90 to 100 per cent. copper.

entirely of β solid solution, in the inner regions which had cooled more slowly the β had partially decomposed with some formation of α solution. The expansion in volume resulting from this structural change had produced internal stresses which led to the failure of the fuse body by cracking.

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APPENDIX C

ALUMINIUM BRASSES

Literature—Effect of Repeated Remelting—Mechanical Properties—
Hot Working—Season Cracking—Corrosion Resistance—Oxidation
at High Temperature—Metallography.

IN the Durville casting process (see Chapter XIV, p. 134) the presence of aluminium in the brass is of great advantage in modifying the character of the molten metal surface. The value of an added element as a feature of the casting method is, however, dependent on its effect on the properties of the alloy. The aluminium brasses not only present a wide range of possible commercial alloys, but the pronounced surface effects produced in the molten brass suggest a related effect in the solid alloys. Consequently the authors carried out an investigation on aluminium brasses,¹ one of the results of which was the discovery of the remarkable combination of properties possessed by the alloys containing 2 per cent. of aluminium; the composition originally selected (76/2/22) for development as a corrosion resistant alloy is now established as a successful material for the manufacture of condenser tubes.

Early investigations on the aluminium brasses were made by Guillet,² who attempted to estimate a value for the zinc equivalent of aluminium. Carpenter and Edwards³ later made a metallographical survey of the copper-rich region of the ternary system, and showed that this coefficient of equivalence was variable, being different for the α and β phases. Reference was also made to the use of aluminium brasses for purposes requiring resistance to corrosion. Records are available of the mechanical qualities of various aluminium brasses, included in publications by Thibaud⁴ and Smalley.⁵ More recently Thews⁶ has dealt with observations on the behaviour and properties of aluminium brasses, especially from the practical foundry standpoint. The advantages of aluminium as a deoxidiser and its effect in minimising the

formation of dross are pointed out. Genders, Reader and Foster,⁷ have given the results of work on the aluminium brasses from the point of view of application to die-casting. Included in that work was a preliminary construction of a portion of the ternary equilibrium diagram. O. W. Ellis,⁸ dealt with complex brasses of about 57 per cent. copper; the high tensile brasses are regarded as aluminium brasses to which iron and manganese are added to increase the ductility and strength respectively. Some results have also been published by the Deutsche Gesellschaft für Metallkunde on the influence of aluminium, lead, iron, and tin on brass.⁹ Tests were made on rolled sheet made from a series of alloys containing 68 and 62 per cent. copper, and the results generally confirm those of Smalley. Considerable irregularities in properties were shown in some of the tests, and were ascribed to flaws possibly originating in the casting process or caused by rolling.

The effect of aluminium in producing a marked decrease in the rate of evolution of zinc vapour from molten brass has been referred to earlier. Pure 70/30 brass at a temperature of 1100° C. evolves copious clouds of zinc fume, but with the addition of as small an amount as 0.1 per cent. of aluminium this evolution is greatly reduced. When the film of aluminium oxide covering the surface is broken, zinc vapour is immediately evolved, but within a few seconds a new film is formed over the disturbed area and evolution of zinc ceases. It is thus evident that the vapour pressure of the zinc is not greatly affected by the presence of aluminium in the alloy, and that the influence of the skin of aluminium oxide is mechanical. The skin on the molten alloy must therefore be under appreciable stress.

The prejudice against aluminium which has commonly existed in brass foundries is based, not only on difficulties connected with casting and an alleged increased susceptibility to season-cracking, but also on the belief that alloys containing aluminium cannot be remelted time after time without deterioration in quality. In the case of pure aluminium, the fear of such deterioration has been disposed of by tests carried out at the National Physical Laboratory.¹⁰

Similar tests have been made with a brass containing 68 per

cent. copper and 2 per cent. aluminium. These consisted in casting, rolling and recasting, repeated until the material had been melted ten times. No flux was used, and no precautions against oxidation were taken during melting. The alloy was poured from the crucible and the rolled strip received no pickling after the intermediate annealing during rolling.

TABLE XVIII

TENSILE TESTS, HARDNESS AND DENSITY
ON BRASSES CONTAINING ALUMINIUM
Test-Pieces Machined from Cast Ingot.

Alloy No.	Analysis.		Density. Gms./c.c.	Max. Proof Stress. Tons/in. ²	Max. Load. Tons/in. ²	Elongation. Per cent. on 2 in.	Reduction of Area. Per cent.	Structure.
	Aluminium. Per cent.	Copper. Per cent.						
1	Nil	70.1	8.50	4.0	14.72	46	41	α
2	0.09	69.8	8.50	4.0	16.28	69	58	α
3	0.25	69.3	8.48	4.8	16.88	69	57	α and trace β
4	0.4	70.1	8.47	4.0	15.64	56	47	α and trace β
5	1.1	68.1	8.36	6.0	20.32	60	59	$\alpha + \beta$
6	2.0	68.0	8.21	8.0	26.48	45	42	$\alpha + \beta$
7	1.9	70.0	8.27	7.0	21.64	58	58	$\alpha + \beta$
8	1.8	74.3	8.33	4.4	17.16	84	74	α
9	2.1	78.1	8.37	4.0	17.64	83	57	α
10	4.7	73.6	7.91	25.0	34.9	23	24	$\alpha + \beta$
11	4.7	76.7	7.93	11.7	31.2	41	36	$\alpha + \beta$
12	4.8	79.7	7.98	9.4	27.0	42	38	$\alpha + \beta$
13	4.6	82.0	8.00	6.5	22.8	65	55	α
14	5.6	74.0	7.80	30.7	40.8	2	7	$\alpha + \beta$
15	5.6	76.2	7.82	16.1	35.2	15	18	$\alpha + \beta$
16	5.7	79.2	7.86	12.1	31.8	33	32	$\alpha + \beta$
17	5.8	81.1	7.87	9.9	27.9	28	26	$\alpha + \beta$
18	5.9	85.0	7.91	6.8	24.0	44	42	α
19	6.8	73.5	7.69	—	24.0	1	3	β and trace α
20	6.8	76.4	7.71	24.4	36.4	2	5	$\alpha + \beta$
21	6.8	79.7	7.74	12.0	34.9	19	36	$\alpha + \beta$
22	7.1	82.3	7.76	9.5	33.7	27	28	$\alpha + \beta$
23	6.9	85.2	7.82	9.4	30.1	44	22	$\alpha + \beta$
24	6.3	88.9	7.95	5.1	20.7	82	57	α
25	7.5	76.3	7.63	6.8	24.4	4	5	β and trace α
26	7.6	79.2	7.64	13.2	30.4	2	5	$\alpha + \beta$
27	7.5	82.3	7.68	9.8	32.9	13	20	$\alpha + \beta$
28	7.4	85.2	7.72	8.6	32.4	31	36	$\alpha + \beta$
29	7.5	87.9	7.75	8.8	30.5	48	44	$\alpha + \beta$

Systematic tests made on the rolled material at each stage showed that repeated remelting has no effect on the properties of aluminium brass, the tensile strength and ductility remaining practically constant throughout. The loss of zinc on melting is remarkably small, unless the metal is overheated, and the aluminium content remains nearly constant.

The properties of aluminium brasses (chill-cast bars 8×1 inch diameter) covering a considerable range of composition are given in Table XVIII. The compositions selected are in six groups, the first showing the effect of increasing aluminium content in brass of about 70/30 composition, and the remainder having in each case constant aluminium contents with varying percentages of copper. Remarkably high ductility is shown by some of the alloys, and in general the aluminium brasses appear mechanically to have distinct advantages. In a number of the alloys the elongation in the tensile test is equal to or higher than that of 70/30 brass, and is accompanied by increased tensile strength.

Table XIX gives the mechanical properties of a number of alloys in the extruded condition.

The effect of aluminium on the hardening of brasses by cold rolling is shown by the results of hardness tests given in Table XX. These tests were made on 9×2×1 inch chill cast

TABLE XIX

TENSILE TESTS ON BRASSES CONTAINING ALUMINIUM

Test-Pieces Machined from Extruded Bar.

Alloy No.	Composition.		Density. Gms./c.c.	Max. Proof Stress. Tons/in. ² .	Max. Load. Tons/in. ² .	Elongation. Per cent. on 2 in.	Reduction of Area. Per cent.	Structure.
	Aluminium. Per cent.	Copper. Per cent.						
30	Nil	69.9	8.524	5.0	21.2	75	70	α
31	"	64.9	8.448	5.7	20.2	62	78	α
32	"	60.0	8.380	6.6	25.7	54	62	α + β
33	2.2	75.7	8.301	5.2	22.7	82	72	α
34	2.2	70.7	8.242	8.6	25.1	66	79	α and trace β
35	2.2	65.8	8.159	10.2	31.2	44	44	α + β
36	3.9	81.1	8.130	5.4	23.7	78	66	α
37	3.9	76.7	8.062	9.8	27.8	54	55	α and trace β
38	3.9	71.9	7.994	10.8	32.2	35	33	α + β

TABLE XX

EFFECT OF ALUMINIUM ON THE HARDENING OF BRASSES BY COLD ROLLING

Hardness after Successive Reductions without Intermediate Annealing.

Analysis.		Brinell Hardness Numbers. $\frac{L}{D^2} = 10$.						
		Percentage Reduction in Thickness.						
Copper. Per cent.	Alumin- ium. Per cent.	Nil %.	20%.	40%.	60%.	80%.	90%.	97%.
70.1	Nil	45	126	140	167	183	186	198
69.8	0.09	48	118	155	160	181	—	—
69.3	0.25	45	109	152	161	183	—	—
70.1	0.4	46	106	135	163	187	—	—
68.1	1.1	61	124	155	167	195	—	—
68.0	2.0	84	156	181	197	208	—	—
70.0	1.9	68	140	165	186	204	205	216
74.3	1.8	54	116	137	174	200	201	212
78.1	2.1	50	121	139	181	195	200	208

ingots of alloys 1 to 9, the compositions and properties of which are given in Table XVIII, and show that the rolling qualities of 70/30 brass are not greatly affected by the presence of aluminium in the percentages used. The ductility of high copper, 2 per cent. aluminium brasses is shown to be extremely high. The behaviour of the aluminium brasses under the very severe test of rolling cold to a reduction of 97 per cent. was considered very satisfactory. Some edge-cracking occurred in the strips, but only beyond a reduction of 80 per cent.

The effect of aluminium on the hot-working properties of the brasses has been investigated by tests similar to those described by Doerinckel and Trockels.¹¹ Compression tests on cylinders (alloys 30-38, Table XIX) carried out at 500°, 600°, 700°, and 800° C. gave the results shown in Fig. 113. These represent the effect of variation in temperature on the work required to compress the cylinders to 50 per cent. of the original length.

In general, the determinations on the pure brasses are in

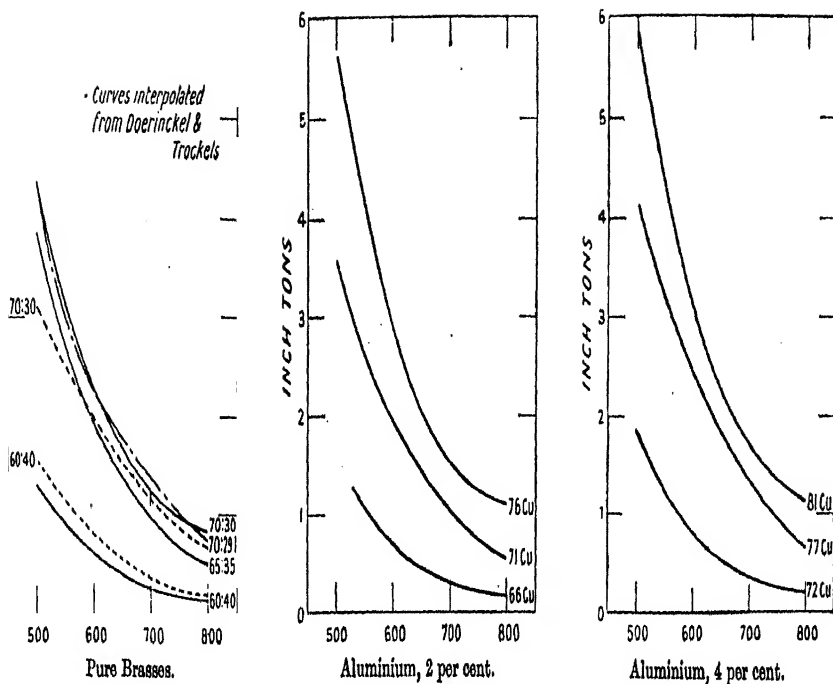


FIG. 113.

Curves showing effect of aluminium on hot working properties of brass.

Work required for compression of a cylinder (0.714 in. diam. \times 1.428 in.) to 50 per cent. of original length at various temperatures.



close agreement with those of Doerinckel and Trockels. The aluminium brasses containing about 2 and 4 per cent. of aluminium, in which the copper contents are raised to 76 per cent. and 81 per cent. respectively (to give a structure entirely of α solid solution), are appreciably harder at 500° C. than 70/30 brass, but with increasing temperature the work required for compression falls rapidly, and at 800° C. there is little difference in the workability of the three alloys. In all the alloys consisting of α constituent up to 800° C., the work required for compression decreases with rising temperature from 500° to 800° C. in the proportions shown below, the different amounts of work required by the three alloys at 500° C. being taken as 100.

	500° C.	600° C.	700° C.	800° C.
α Pure brass	100	55	29	19
α 2 per cent. aluminium brass	100	51	26	19
α 4 per cent. aluminium brass	100	53	30	19

Aluminium brasses of lower copper content show a behaviour similar to that of pure brasses of similar constitution over the range of temperatures used. Aluminium brasses of 71 per cent. copper with 2 per cent. aluminium, and 77 per cent. copper with 4 per cent. aluminium are similar to 65/35 pure brass; those of 66 per cent. copper with 2 per cent. aluminium, and 72 per cent. copper with 4 per cent. aluminium are similar to 60/40 pure brass. In each series of alloys the alloy which has a structure consisting entirely of β at 800° C. requires only about one-sixth of the work necessary to produce the same effect in an alloy consisting of α at that temperature.

Although there appear to be no published data in connection with the season-cracking of aluminium brasses, the opinion has existed that aluminium increases the susceptibility of brasses to this defect. The results of an investigation carried out by the authors do not confirm this view.

Three alloys were used, having the compositions given below :

Copper. Per cent.	Aluminium. Per cent.	Zinc. Per cent.
76.26	2.10	Remainder
69.11	Trace	"
68.45	Nil	"

For tests using externally applied stress, flat test-pieces were machined from the strips cold rolled to various hardnesses. After determining the breaking load of each type of strip from each alloy, test-pieces were pickled and subjected in the testing machine to loads below the breaking load, while the surface of each test-piece was coated with mercury by treatment with mercurous nitrate solution.¹² The time required for cracking to occur was determined. For tests on the effect of internal stress, small spun cups were coated with mercury or exposed to the action of ammonia. The results indicated clearly that aluminium does not increase the susceptibility of brass to season-cracking. In all cases there was no appreciable distinction between the aluminium brasses and the brass free from aluminium.

Corrosion tests have shown that the resistance of brasses to corrosion is considerably improved by the addition of aluminium. Aluminium brasses of high copper content (α brasses) are superior to those of lower copper content (β brasses),* and definitely more resistant than brasses free from aluminium.

In view of the results obtained by the authors, further work on the corrosion of aluminium brass was taken up in the investigations of the Corrosion Research Committee of the Institute of Metals at the laboratory of the Royal School of Mines, London.

The alloy used, containing 76 per cent. copper, 2 per cent. aluminium and 22 per cent. zinc, was selected as having mechanical properties similar to those of 70/30 brass, combined with the minimum percentage of aluminium capable of giving high resistance to corrosion and oxidation at high temperatures.†

* It has been observed also in the molten alloys that the film of aluminium oxide appears to have greater strength in the high-copper brasses than in those of lower copper content.

† British Patent 308,647.

This material proved to have remarkable resistance to "impingement attack" in the tests developed by R. May; the results are included in the Eighth Report to the Corrosion Research Committee of the Institute of Metals.¹³

An outstanding property of the aluminium brass is its capacity for forming a self-healing surface film under corrosive conditions, and the alloy has now become a regular commercial product for the manufacture of condenser tubes, etc.

On storage under indoor conditions, brasses containing aluminium resist surface corrosion more efficiently than pure brasses, whether the surface is that of the original casting or has been prepared by polishing or etching.

One of the most interesting features shown by a comparison of the behaviour of different alloys is that those containing aluminium appear to be remarkably resistant to oxidation both at ordinary and elevated temperatures. An ingot or die-casting of aluminium-bronze or brass containing aluminium invariably shows a clean piping depression, and even when removed from the mould immediately after casting possesses a clean soft metallic lustre on cooling to atmospheric temperature. Effective use of this property has been made during research on the macrostructure of cast metals, especially in connection with the preparation of "bled" ingots, in which the form of the cavity left by liquid metal drained from the partly solid mass is of considerable importance. Cavities prepared in this way have remained untarnished for several months without the application of protective materials.

Investigation has shown that, where the surface advantages conferred by aluminium are required, such as in die-casting or the type of investigatory work described, quite a small addition (0.1 per cent. or less), insufficient to modify appreciably the properties of the alloy, is adequate.

The surface of aluminium brass poured while exposed to the air shows no indication of oxidation of copper, and since the formation of copper oxide is an essential condition in the production of "red stain" in the pickling of brass, it would be inferred that the addition of aluminium to brass may provide a method of avoiding the occurrence of "red stain" without recourse to special annealing or pickling processes. The surface oxidation of aluminium brasses has been studied by

heating rolled 70/30 brass, containing 0, 0.1, 0.2, 0.4, 1, and 2 per cent. of aluminium, in an oxidising atmosphere at 650° C. At this temperature oxidation occurs in each case, but there is a marked difference between the brasses containing less

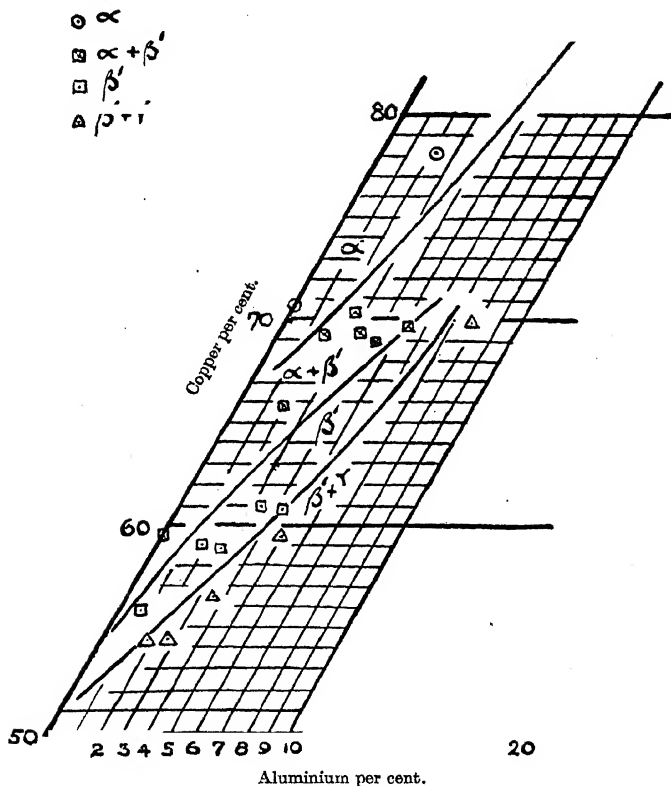


FIG. 114.

Structure of chill cast aluminium brasses.

than 1 per cent. aluminium, which become thickly oxidised and give red stains on pickling, and those containing 1 and 2 per cent. aluminium, which oxidise only sufficiently to dull the surface and are bright and free from red stain after pickling. The character of the oxide formed depends to a great extent on the nature of the original surface; the beneficial effect

of aluminium is shown to the fullest extent on rolled and polished strip.

In a further development of this work, included in an exhaustive research on the oxidation of brass, J. S. Dunn¹⁴ has determined quantitatively the rate of oxidation of alumin-

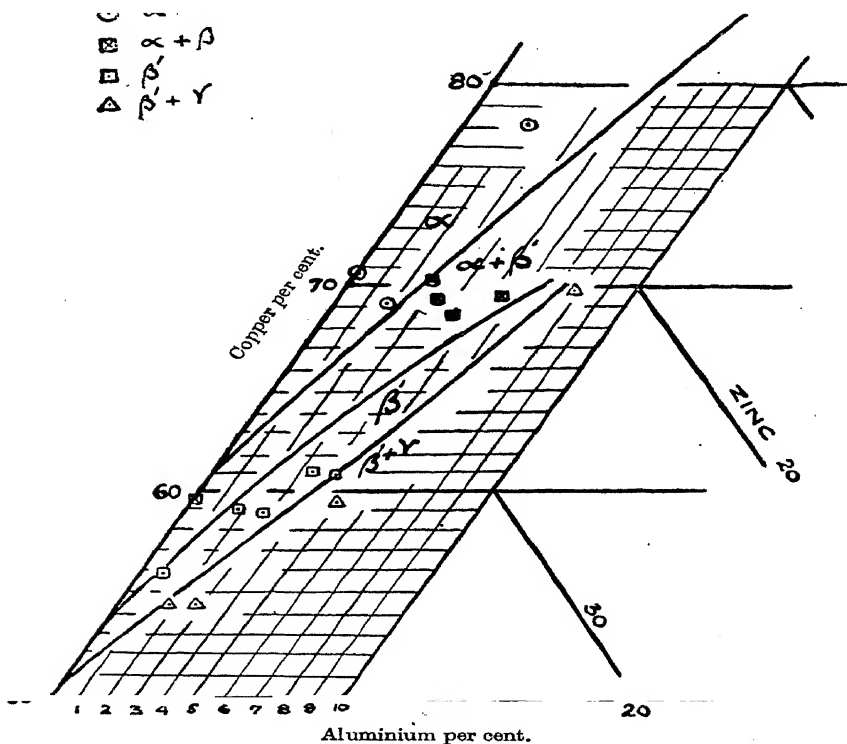


FIG. 115.

Structure of annealed aluminium brasses.

ium brasses. His results confirm the remarkable difference in the behaviour of pure brasses and aluminium brasses when exposed to oxygen at high temperatures; the rate of oxidation of brass containing 2 per cent. aluminium is only about one-fortieth of that of pure 70/30 brass.

The microstructures of the aluminium brasses, indicated in the last column of Tables XVIII and XIX, are generally

similar to those of pure brasses of the same phase constitution. Approximate phase diagrams covering the range of alloys dealt with are given in Figs. 114 and 115, constructed from micrographic evidence from the alloys as cast and in the worked and annealed conditions. The results of this preliminary study are in close agreement with those of the later complete investigation by Bauer and Hansen.¹⁵

The observations described show that, from several aspects, marked practical advantages are secured by the use of aluminium as an addition to the ductile brasses. The beneficial effects are due both to the influence of aluminium as a metallic constituent of the alloy and to the formation of a protective surface film of aluminium oxide.

The only difficulty in the way of taking full advantage of the properties conferred on brass by aluminium lies in the casting process. In foundries a strong prejudice against the use of aluminium has long existed, and this is undoubtedly based on sound practical observation. The results now being obtained in commercial practice by the Durville casting method show, however, that difficulties attendant on the use of aluminium are surmountable.

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APPENDIX D

PHOSPHORUS BRASSES

Mechanical Properties—Structure.

THE use of additions of special elements to brasses as a means of modifying the casting characteristics of the alloys involves in each case an examination of the manner in which the properties of the alloy and its behaviour in working are affected. This is necessary to determine the ultimate value of the method, the limiting amounts of the element to be added and any required changes in treatment or composition. Especially is this the case with phosphorus, which is known to exert a pronounced effect when present in small quantities in other copper alloys.

Previous work on the effect of phosphorus in brass is described in the 11th Report of the Centralstelle¹ and by A. Portevin.² The data given show that additions of phosphorus in amounts of the order of 0.05 per cent. have only a slight influence on the hardness of pure 70/30 brass. The presence of iron in conjunction with phosphorus, however, results in a considerable increase in hardness, as shown below in Table XXI, extracted from the 11th Report of the Centralstelle.

TABLE XXI

EFFECT OF IRON ON HARDNESS OF PHOSPHORUS BRASS

Composition of Brass.	Iron Content %.	Brinell Hardness.
Copper 72% Phosphorus 0.04% Annealed 700° C. }	Nil	54
	0.02	68
	0.04	74
	0.07	78

In view of the advantages of phosphorus addition in modifying the surface characteristics of molten brass, the authors have made a fuller investigation of the mechanical properties and structural features of the phosphorus brasses.

The results (Table XXII) show that the influence of phosphorus up to 0.1 per cent. is small in cast 70/30 brass and that no appreciable effect is produced by annealing. Increase in phosphorus content to 0.4 per cent. raises the hardness and decreases considerably the ductility of the alloy. Ingots containing phosphorus up to 0.4 per cent. can, however, be cold rolled to strip without difficulty by the usual commercial practice, including reductions of 50 per cent. in thickness before annealing. The results of tensile tests made on various phosphorus brasses in the rolled condition and after annealing are given in Table XXIII. Only the brass containing 0.41 per cent. phosphorus showed distinctly greater work hardening than 70/30 brass.

The relative rates of hardening during cold rolling of ingots of pure brass and brass containing 0.04 per cent. phosphorus, both in the "as cast" and annealed conditions, are shown in Table XXIV, which gives the Brinell hardness of specimens rolled without intermediate annealing to successive stages of 23, 41, 59, 68, 78, 89 and 94 per cent. reduction in thickness. The rate of work-hardening of the low phosphorus brass at all stages of rolling is almost identical with that of pure 70/30 brass. Annealing of the ingot before rolling produces no sensible effect on the behaviour of the materials. That the working properties of brass containing 0.04 per cent. phosphorus are generally satisfactory is shown by the fact that small cartridge cases can be made from the material by the same methods of drawing and intermediate annealings as are used for pure 70/30 brass, the only difference between the two materials being a slightly higher final hardness of the finished cases made from phosphorus brass. The results of Brinell hardness determinations made at various positions on finished cases of phosphorus brass (0.04 per cent. phosphorus) and 70/30 brass are given below.

Brinell Hardness.

Ins. from Flange	0.15.	0.20.	0.30.	0.70.	1.10.	1.50.
Phosphorus brass	151	151	158	171	161	114
70/30 brass	142	134	146	153	153	100

TABLE XXII
TENSILE TESTS ON CAST PHOSPHORUS BRASSES

Composition Per Cent.			Test Piece.	Yield Point. Tons/sq. in.	Ultimate Strength. Tons/sq. in.	Elonga- tion. %.	Reduction of Area. %.	Brinell Hardness.
Copper.	Phosphorus.	Iron.						
71.1	0.41	0.01	As cast	5.8	18.9	44	—	63
			Annealed	5.7	16.3	33	28	63
70.7	0.09	0.03	As cast	4.4	17.9	68	54	55
			Annealed	5.8	17.4	58	43	53
70.9	0.045	0.02	As cast	4.7	16.4	72	63	54
			Annealed	4.2	16.6	76	53	55
72.2	Nil	0.02	As cast	4.8	17.3	80	64	49
			Annealed	4.3	17.0	72	54	47

TABLE XXIII

PHOSPHORUS BRASSES. PROPERTIES OF ROLLED STRIP

Analysis Per cent.				Strip 0.5 in. thick, 50% Reduction.				Strip 0.25 in. thick, 75% Reduction.				Strip 0.125 in. thick, 87.5% Reduction.				Strip 0.06 in. thick, 94% Reduction.
Cop- per.	Phos- phorus.	Iron.	Test Piece.	Tensile Strength, Tons/in.*	Elonga- tion, %.	Reduc- tion of Area, %.	Brinell Hard- ness.	Tensile Strength, Tons/in.*	Elonga- tion, %.	Reduc- tion of Area, %.	Brinell Hard- ness.	Tensile Strength, Tons/in.*	Elonga- tion, %.	Reduc- tion of Area, %.	Brinell Hard- ness.	Brinell Hardness.
71.1	0.41	0.01	Rolled Annealed	40.7 25.0	1 60	2 50	182	45.7 24.4	6 72	30 63	191	44.2 24.3	10 81	32 60	192	196
70.7	0.09	0.03	Rolled Annealed	36.4 23.9	12 64	32 88	185	42.5 23.5	12 66	39 70	188	40 23.0	11 70	44 70	181	191
70.9	0.045	0.02	Rolled Annealed	35.0 22.0	15 68	44 73	182	38.8 20.6	14 80	51 76	178	36.8 20.6	14 82	33 75	185	183
72.2	Nil	0.02	Rolled Annealed	34.5 20.3	15 76	43 76	163	33.3 20.1	16 80	51 72	177	33.4 20.5	15 88	42 67	168	176

TABLE XXIV

EFFECT OF 0.04 PER CENT. PHOSPHORUS ON THE RATE OF HARDENING OF 70/30 BRASS

Material.		Brinell Hardness figures at various Stages of Cold Rolling.							
Condition.	Composition.	Reduction Nil.	23%.	41%.	59%.	68%.	78%.	89%.	94%.
As cast	70.4% Cu, Phosphorus Nil	57	112	151	168	182	190	199	207
	70.9% Cu, 0.04% P, 0.02% Fe	60	107	157	175	184	193	203	210
Cast and Annealed for 1 hour at 650° C.	70.4% Cu, Phosphorus Nil	59	104	146	172	184	190	197	206
	70.9% Cu, 0.04% P, 0.02% Fe	60	110	152	168	176	190	201	209

The effect of phosphorus on the hardness of brass annealed in the cast state and after rolling is shown by the results of tests given in Table XXV. The presence of phosphorus gives an appreciably higher hardness after annealing, and the values obtained indicate that an increase in annealing temperature of about 50° C. is necessary to reduce the hardness of phosphorus brass to that of pure brass annealed at 650° C. Higher temperatures produce additional softening.

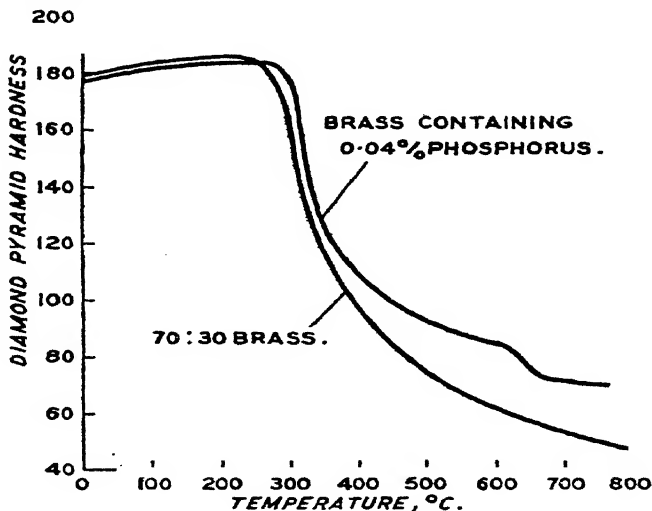


FIG. 116.

Effect of annealing on hardness of brasses with and without phosphorus.
(Cooke and Miller³).

Some recent work by Cook and Miller³ on the effect of different elements on the annealing and grain growth of α brass included brasses containing phosphorus, and the results are in general agreement with the above; Fig. 116 shows curves extracted from the paper by Cook and Miller, illustrating the effect of annealing on the hardness of brasses of 70/30 composition with and without the addition of phosphorus. The deflection of the curve of the phosphorus brass after the commencement of softening and recrystallisation is characteristic also of brass containing certain other added elements. This indicates a retardation of softening, the effect of some opposing factor possibly akin to precipitation hardening.

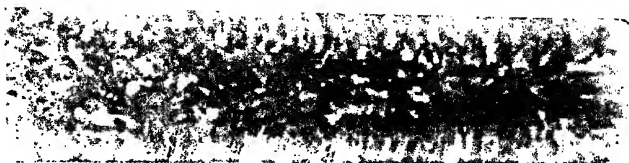
TABLE XXV

EFFECT OF PHOSPHORUS ON THE HARDNESS OF
STRIP BRASS

Material.	Per- centage Reduction by Cold Rolling.	Brinell Hardness.				
		As Rolled.	Annealed 1 hour at			
			600° C.	650° C.	700° C.	750° C.
70/30	41	151	63	61	—	—
Chill Cast	68	182	59	59	—	—
Phosphorus	41	157	76	71	—	—
Brass	59	175	—	—	58	51
Chill Cast	68	184	73	73	—	—
(P 0.04%, Fe 0.02%)						
70/30	41	146	59	58	—	—
Chill Cast and	68	184	56	56	—	—
annealed 1 hour						
at 650° C.						
Phosphorus	41	152	74	72	—	—
Brass (P 0.04%,	68	176	71	71	—	—
Fe 0.02%)						
Chill Cast and						
annealed 1 hour						
at 650° C.						

Ingot sections of 70/30 brass containing 0.04 per cent. phosphorus, chill cast with an inert mould coating of soot, show the same type of structure as that found in ingots of pure 70/30 brass cast under similar conditions, consisting of a central zone of equiaxial crystals and a columnar outer zone extending to the surface. Ingots cast in a water-cooled copper mould consist almost entirely of fine columnar crystals, except near the top of the ingot. Figs. 117 and 118 show etched sections of typical ingots of phosphorus brass. The unsoundness, measured by density determinations on whole ingots, is of the same order as that found in 70/30 brass ingots cast under similar conditions but with a protective reducing flame.

The effect of the presence of phosphorus on the micro-structure of cast brass is to produce intensified "coring" of the crystals and the appearance of free phosphide. With a



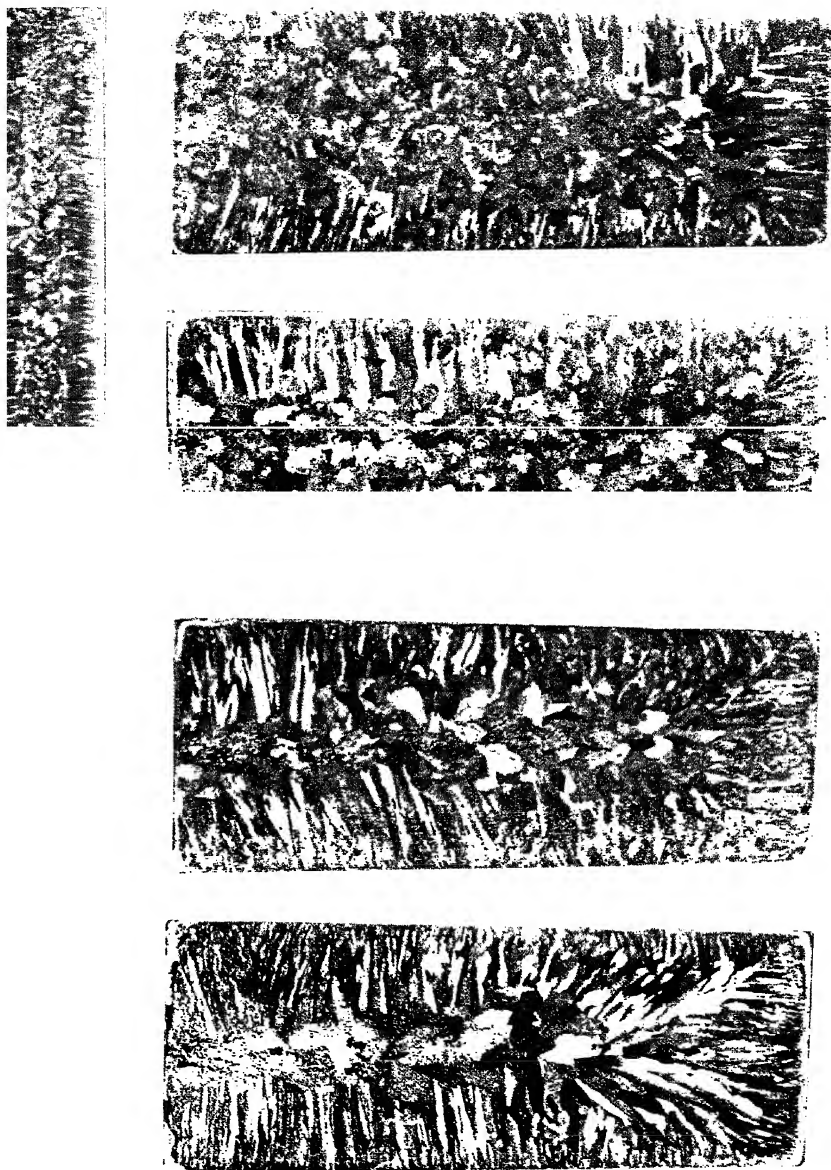
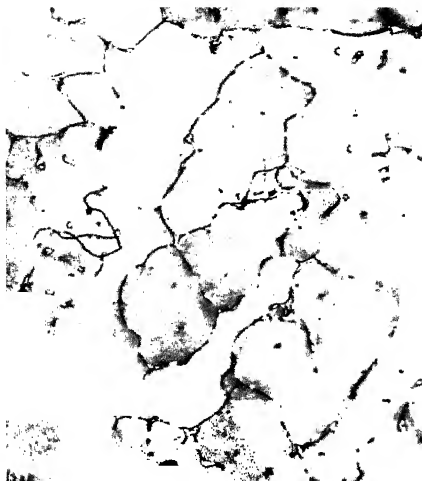
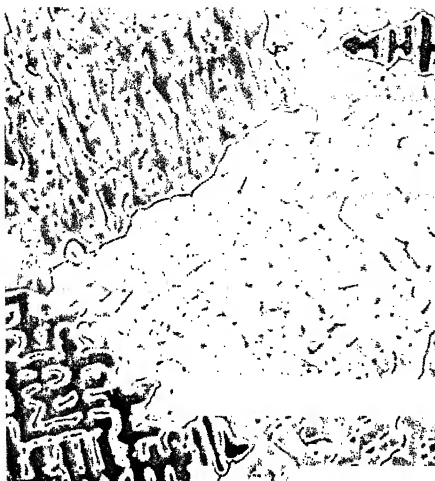


FIG. 118.

Etched sections of $24 \times 6 \times 1.5$ inch phosphorus brass ingot cast in solid copper mould.



(b)

FIG. 119.

Microstructure of 70/30 brass containing 0.41 per cent. phosphorus. $\times 100$.

(a) As cast.

(b) Annealed 650° C. for one hour.

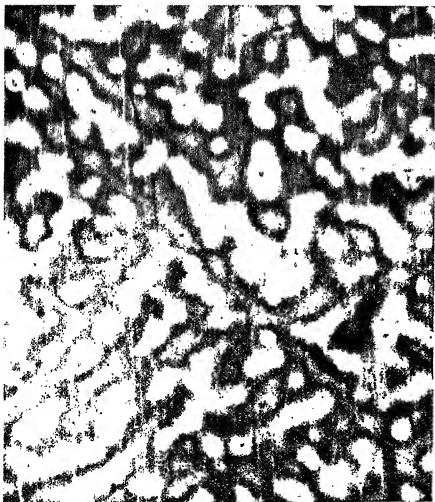


FIG. 120.

Microstructure of 70/30 brass containing 0.09 per cent. phosphorus. As cast. $\times 100$.



FIG. 121.

Microstructure of 70/30 brass containing 0.045 per cent. phosphorus. As cast. $\times 100$.



(a)



(b)

FIG. 122.

Effect of phosphorus on grain size of 70/30 brass cold rolled and annealed 600°C . for one hour. $\times 100$.

(a) Pure 70/30 brass.

(b) 70/30 brass plus 0.04 per cent. phosphorus.



(a)



(b)

FIG. 123.

Effect of phosphorus on grain size of 70/30 brass cold rolled and annealed 650°C . for one hour. $\times 100$.

(a) Pure 70/30.

(b) 70/30 brass plus 0.04 per cent. phosphorus.

phosphorus content of 0.045 per cent. the amount of free phosphide is very small, occurring as thin films. Increase of phosphorus to 0.41 per cent. produces a marked increase in the number and size of the phosphide areas. Representative structures of cast 70/30 brass containing 0.41 per cent., 0.09 and 0.045 per cent. phosphorus are shown in Figs. 119, 120, 121. The effect of annealing the cast material is to reduce the "coring" effect by diffusion, and in brass containing 0.045 per cent. phosphorus, the whole of the free phosphide is taken into solution by annealing at 650° C. for one hour, the structure then consisting of clear α crystals similar to pure 70/30 brass. With 0.09 per cent. phosphorus the free phosphide is nearly but not completely removed by a similar annealing, while with 0.41 per cent. phosphorus considerable quantities remain, mainly in the form of continuous films.

Examination of worked and annealed material shows that, compared with pure 70/30 brass under identical conditions, phosphorus brass invariably possesses a finer structure. The structures of 70/30 brass free from phosphorus and with addition of 0.04 per cent. phosphorus, annealed at 600° C. for one hour after different stages of cold rolling are shown in Figs. 122 and 123. The smaller grain size of the phosphorus brass corresponds to its higher hardness.

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APPENDIX E

LIST OF REPORTS AND NOTES ON THE CASTING OF BRASS INGOTS

Communicated as confidential reports to members of the British
Non-Ferrous Metals Research Association.

B.N.F.M.R.A. Research Report No.	Investi- gators' Report, Part No.	Author.	Subject.
12	I.	Genders, R.	Survey of present practice.
21A	II.	Genders, R. Bailey, G. L. Haughton, M. A.	Examination of material of cur- rent manufacture, the occur- rence, characteristics and ori- gin of defects.
60	IIA.	Genders, R. Bailey, G. L. Haughton, M. A.	Examination of further material of current manu- facture.
25A	III.	Moore, H.	The Durville casting process.
44	IIIA.	Genders, R. Haughton, M. A.	Examination of ingots cast by the Durville process.
45	IV.	Genders, R. Bailey, G. L.	Taper ingots.
46	V.	Genders, R. Bailey, G. L.	The influence of casting tem- perature and rate of pouring on soundness and surface quality of the ingot.
67	VI.	Genders, R. Bailey, G. L.	The influence of feeding on the soundness of top-poured in- gots.
68	VII.	Genders, R.	Directional solidification.
65	VIII.	Genders, R. Bailey, G. L.	The density of the brasses. (See also <i>J. Inst. Metals</i> , 1925, 33, 191.)
77	IX.	Genders, R.	Possible advances in casting methods.
91	X.	Genders, R. Bailey, G. L.	Effect of large ingot thick- ness.
106	XI.	Genders, R.	Use of aluminium as an ad- dition to brass alloys. (See also <i>J. Inst. Metals</i> , 1930, 43, 163.)
195	XII.	Genders, R. Bailey, G. L. Northcott, L.	Surface and subsurface sound- ness of ingots.
142	XIII.	Bailey, G. L.	The influence of mould con- ditions on the solidification of 70/30 brass ingots.

B.N.F.M.R.A. Research Report No.	Invest gators' Report, Part No.	Author.	Subject.
147A	XIV.	Bailey, G. L.	The influence of dissolved gases on the soundness of 70/30 brass ingots. (See also <i>J. Inst. Metals</i> , 1928, 29, 191.)
138	XV.	Genders, R. Northcott, L.	The centrifugal casting of 70/30 brass.
175	XVI.	Bailey, G. L.	Examination of 70/30 brass ingot cast in Junker water-cooled mould.
191	XVII.	Genders, R. Bailey, G. L.	The effect of aluminium on the hot working properties of the brasses.
193	XVIII.	Bailey, G. L.	Mould materials. (See also <i>J. Inst. Metals</i> , 1932, 49, 203.)
194	XIX.	Genders, R. Bailey, G. L.	Large-scale trial of brass cast by the R.D. modified method.
282	XX.	Genders, R. Northcott, L.	Phosphorus brasses.
228	XXI.	Genders, R. Bailey, G. L.	Summary report of the research.
316	XXII.	Northcott, L.	Composition of flame gases and mould atmospheres using different types of gas burner.
	XXIII.	Northcott, L.	The occurrence of blisters in cross-rolled brass sheet.
	Interim Note No.		
153	1	—	The rotatory (Durville) process of casting. (Effect of pouring conditions on structure, etc.)
154	2	—	The effect of rolling conditions on the quality of brass strip.
—	3	—	A simplification of the modified method of casting.
155			Results of tests on sample ingots of aluminium brass and aluminium nickel silver supplied by manufacturers.
			Manufacture of aluminium brass tubes by the extrusion process.
200	6		Bottom casting.
—	7		Examination of samples of aluminium brass condenser tubes of current manufacture.
239	8		Resistance to fatigue of the aluminium brasses.
309	13		Examination of a large blister defect in brass strip.
125	—		Notes on apparatus used for the production of ingots by an improved casting method.
B.N.F.M.R.A. Development Report No.			
11	—		Erichsen water-cooled moulds. A description of the "Erichal"

B.N.F.M.R.A.
Development
Report No.

Subject.

			mould and the results of the examination of an " Erical " cast brass ingot.
16	—	—	The casting of strip brass ingots. Modified casting method: modification of the gas burner for use with low-pressure gas.
17	—	—	Abnormal hot-brittleness in 70/30 brass sheet.
—	—	—	The formation of blisters in cross-rolled brass sheet.

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